



On the Short Surface Fatigue Crack Growth Behavior in a Fine-Grained WC-Co Cemented Carbide

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Abstract: In the present study, the fatigue crack growth (FCG) behavior of short surface cracks in a fine-grained cemented carbide with a length of less than 1 mm was investigated. The rotating bending and the four-point bending fatigue tests were carried out at stress ratios of R = -1 and R = 0.1 (R = maximum stress/minimum stress). It was found that a short surface crack had a longer stable fatigue crack growth area than a long through-thickness crack; the FCG behaviors of the two types of crack are clearly different. Furthermore, the FCG path of short surface cracks was investigated in detail to study the interaction between fatigue cracks and microstructures of the cemented carbide such as WC grains and the Co phase. At a low K_{max} (K_{max} = the maximum stress intensity factor), it was found that fatigue crack growth within WC grains is difficult because of a small driving force; instead, crack growth is along the brittle WC/WC interface. On the other hand, at a high K_{max} , WC grain breakage often occurs, since the driving force of FCG is large, and the fatigue crack grows linearly.

Keywords: WC-Co cemented carbide; short surface fatigue crack; long through-thickness fatigue crack; fatigue crack growth (FCG); crack growth path; stress intensity factor (SIF)

1. Introduction

Cemented carbides are widely used for cutting tools, dies, mechanical parts, and so forth [1], because of their superior abrasive wear resistance to tool steels. Among the cemented carbides, WC-Co alloys have superior mechanical properties and are highly versatile and widely used [2]. The mechanical properties of cemented carbides are intermediate between those of brittle materials, such as ceramics and cermets, and ductile materials such as tool steels. In general, the extent of the brittleness increases as the Co content decreases and the WC grain size decreases. More specifically, as the Co content increases, the toughness of the material increases, but the hardness decreases [3].

There have recently been increasing cases of utilizing cemented carbides for metal dies and increasing demand for long-life dies. When a cemented carbide is applied to a metal die, it may become useless after long-term use due to the fatigue crack growth (FCG). It is desirable to use WC-Co dies safely for long durations without fracture failures such as chipping, cracking, or rupture. Therefore, it is extremely important to understand the relationship between the FCG behavior of cemented carbides and their microstructures such as WC grain and the Co phase [4].



Previous studies on the fatigue behavior of cemented carbides can be broadly classified into (1) determination of the fatigue lifetime and fatigue limit, and (2) evaluation of the FCG behavior based on fracture mechanics or damage tolerance mechanics.

There have been numerous studies in the former category. For example, many papers have reported that the fatigue strength of cemented carbide is decreased by applying a cyclic load [5]. In addition, the reduction of fatigue strength becomes marked with increasing Co content in the material [6–11]. In recent years, studies on the FCG behavior in the latter category have also been conducted, although many of them examined only the behavior of long through-thickness fatigue cracks.

Fry et al. [12] investigated the relationship between the rate of FCG, da/dN, and the stress intensity factor range, ΔK . They reported that the FCG resistance increases with increasing Co content, and with increasing WC grain size. Torres et al. [13] and Llanes et al. [14] studied the FCG behavior of long through-thickness cracks in cemented carbides. They clarified that the rate of FCG more strongly depends on the maximum stress intensity factor K_{max} than the stress intensity factor range ΔK .

Boo et al. [15] investigated the FCG behavior of long through-thickness cracks in a cemented carbide. They reported that the Co phase at the crack tip undergoes a phase transformation from FCC to HCP when subjected to repeated loading, resulting in the embrittlement of the area and accelerated FCG. Sunouchi et al. [16] studied the FCG mechanism of long through-thickness cracks using the acoustic emission method. They showed that the lower FCG resistance of the cemented carbide at a high K_{max} than that in steel and in aluminum alloys is due to transgranular cracking of WC grains. In addition, Mingard et al. [17] observed the FCG path of a long through-thickness fatigue crack in situ, using a scanning electron microscope (SEM) and a specially designed flat specimen [18].

The FCG behavior and the crack paths described above were for long through-thickness fatigue cracks. Excluding the following studies, there have been very few studies on the FCG behavior of short surface fatigue cracks, which often cause the surface cracking of dies made of cemented carbides.

Ishihara et al. [19] conducted four-point bending fatigue tests using rectangular specimens and investigated the FCG behavior of short surface fatigue cracks in a cemented carbide. They found that at a low K_{max} , the cyclic stress loosens the interface between the WC grains and the binder phases, leading to the occurrence of microcracks at the interface in front of the crack tip. Repeated stress also eliminates bridging portions, located behind the crack tip.

When the crack driving force, K_{max} , is sufficiently large to break WC grains, the crack growth process is mainly dominated by the transgranular cracking fracture of WC grains.

Ishihara et al. [20] also investigated the FCG behavior of short surface cracks in cermet, and found that its FCG behavior was similar to that of a cemented carbide. Mikado et al. [21] conducted a rotating bending fatigue test on a cemented carbide to evaluate its fatigue strength and the FCG behavior of short surface cracks in the material. It was found that reducing the rate of FCG at the initial stage of the fatigue fracture process is important for prolonging the fatigue lifetime.

As reviewed above, although the FCG behavior of long through-thickness cracks in cemented carbides has been studied, there has been limited research on the FCG behavior of short surface cracks and the crack growth pathway. This is because the observation of short surface fatigue cracks with a length on the order of 100 μ m in hard brittle materials is extremely difficult compared with the observation of long through-thickness fatigue cracks. In addition, there has been no comparison of the difference between the FCG of short surface cracks and long through-thickness cracks.

In the present study, the FCG behavior of short surface cracks less than 1 mm in length in a cemented carbide were investigated for the first time. Then, a comparison was made between the FCG behavior of long through-thickness cracks and short surface cracks to clarify the difference between them. Furthermore, the crack growth pathway of short surface fatigue cracks was investigated in detail to study the interaction between fatigue cracks and the material microstructure, WC grains, and the Co phase.

2. Materials and Methods

2.1. Material

In the present study, a commercially available fine-grained WC-Co cemented carbide was used as the testing material. Tables 1 and 2 show the chemical composition and mechanical properties of the material, respectively. The material contains 13.0 wt % Co, and its bending strength is 4100 MPa. The Young's modulus was obtained from the catalog. The fracture toughness was the average value of five measurements by the single-edge-precracked-beam (SEPB) method [22]. The bending strength was the average value of three measurements by the three-point bending test using plane specimens as shown in the next section (Figure 3a). The Vickers hardness was the average value of ten measurements performed with a Vickers hardness tester (FUTURE-TECH CORP., Kanagawa, Japan) under a load of 2.94 N and a holding time of 20 s. Figure 1 shows the microstructure of the material, which is a composite microstructure consisting of both a WC and a Co phase. Figure 2a,b shows histograms of the WC grain size and binder mean free path, respectively, constructed from the observation of the microstructure shown in Figure 1. The material has a log-normal distribution of WC grain size with a modal value of about 0.35 μ m, and has an exponential distributed binder mean free path ranging from 0 to 1 μ m.

Table 1. Chemical composition of the WC-Co cemented carbide (wt %).

Со	Cr	W and C
13.0	0.51	Bal.

Young's Modulus (GPa) Fracture Toughness (MPam ^{1/2})		Bending Strength (MPa)	Vickers Hardness (HV)
550	12.1 (11.8–12.6)	4100 (3800–4350)	1480 (1410–1520)

Table 2. Mechanical properties of the WC-Co cemented carbide.



Figure 1. Microstructure of the WC-Co cemented carbide used in the study. The average WC grain size is approximately $0.45 \mu m$.



Figure 2. Histograms of: (a) WC grain size; (b) Binder mean free path.

2.2. Specimens

The shapes and dimensions of the specimens used in the present study are shown in Figure 3. Two types of hourglass-shaped specimens with a minimum diameter of 3 mm were used: the un-notched specimen shown in Figure 3a, and the notched specimen shown in Figure 3b. The former was used to investigate the *S*-*N* curve of the material at a stress ratio of R = -1, while the latter was used to study the FCG rate of short surface cracks in the material at R = -1. Specimens of both types were polished to a mirror-like finish with diamond paste (particle size of 1 µm) to minimize the effects of roughness and residual stress. The surface roughness R_a of the polished specimens was less than 0.1 µm. The compressive residual stress σ_r on the surface of the polished specimens was less than 500 MPa in the WC phase and less than 200 MPa in the Co phase. For the notched specimen, an artificial notch with a length of 0.08 mm, a width of 0.02 mm, and a depth of 0.04 mm was introduced into the specimen surface by laser beam machining. In addition, four-point bending fatigue tests were conducted at R = 0.1 using the rectangular specimens. Figure 4 shows the shape and dimensions of the specimen, which were 8 mm wide, 4 mm high, and 50 mm long. A Vickers indentation (indentation load: 400 N and indentation time 20 s.) was introduced on the center of the specimens to make a pre-crack.



Figure 3. Specimens used for *S*-*N* curve and investigating the fatigue crack growth (FCG) rate of short surface cracks at a stress ratio of R = -1: (a) Plane specimen (un-notched specimen) for determination of *S*-*N* curve; (b) Notched specimen for investigating the FCG rate of short surface cracks.



Figure 4. Notched specimens used for investigating the FCG rate of short surface cracks at R = 0.1.

2.3. Experimental Procedure for Fatigue Tests

2.3.1. Rotating Bending Fatigue Tests (R = -1)

Rotating bending fatigue tests were conducted in air at room temperature at a stress ratio of R = -1 with a stress frequency of 10–15 Hz. The un-notched and notched specimens shown in Figure 3a,b were respectively used to evaluate the fatigue lifetime and the rate of FCG for short surface cracks. The fatigue tests on the notched specimen were interrupted periodically to collect specimen-surface replicas. Acetylcellulose films and acetone were utilized for the replica films and the solvent, respectively. The lengths of fatigue cracks transcribed on these replica films were measured using a laser microscope. Two types of FCG tests were conducted: FCG tests under constant stress amplitudes (increasing- K_{max} tests), and FCG tests at a constant K_{max} . The rate of FCG, da/dN, can be obtained from the plot of 2a versus N, where 2a and N are the crack length and the number of stress cycles, respectively. For the constant stress amplitude tests, the values of da/dN were determined as the slope of the approximate fitting curve to the experimental data at several measurement points. On the other hand, for the constant K_{max} tests, the values of da/dN were determined as the slope of the approximate fitted to the experimental data at each K_{max} value.

Figure 5a shows the specimen surface after the laser beam machining. In this figure, discoloration due to the effect of laser heating can be seen around the notch. It was confirmed that there was some melting of the material as shown in Figure 5b and some cobalt oxidation as shown in Figure 5c in the discolored region. The heat-affected zone was determined as the discolored region. The total length of the discolored region is about 120 to 140 μ m. Figure 5d shows the appearance of a short surface fatigue crack that was initiated and grew from an artificial notch. To eliminate the effect of the heat-affected zone, the region used to measure the rate of FCG was set outside the heat-affected zone.

On the driving force of fatigue cracks in brittle and microscopically heterogeneous materials, a common understanding has not yet been established. In the present study, the maximum stress intensity factor, K_{max} , was utilized, as in previous studies [4,21]. To calculate the K_{max} , the following expression was used:

$$K_{\max} = Y \sigma_{\max} \sqrt{\pi a},\tag{1}$$

where σ_{max} is the maximum stress, *a* is a half the crack length, and *Y* is the crack shape correction factor. A value of 0.73 was used for *Y* assuming that the crack is semicircular [23]. An actual crack does not grow along a straight path, but along a zigzag path because it usually grows along a WC/Co boundary as shown later. The crack length projected in the stress loading direction was employed as the crack length 2*a* in this study.

To investigate the crack growth path in detail, FCG tests were conducted at several constant values of K_{max} using the notched specimen in Figure 3b. The stress amplitude σ_a was adjusted to an appropriate value to maintain a constant value of K_{max} during the fatigue tests, since K_{max} increases with the crack length under a constant stress amplitude. The crack growth path was observed on the specimen surface until the failure of the specimen. After the specimen failure, the fracture surface was also investigated in detail using an SEM to reveal the crack growth path on the fracture surface.



Figure 5. Example of an artificial notch (**a**–**c**) and a short surface fatigue crack initiated from the notch (**d**): (**a**) Optical microscope image of laser beam machined notch; (**b**) SEM image of A around the notch in Figure 5a; (**c**) The oxygen signal mapping in the exact area of Figure 5b evaluated by energy dispersive X-ray spectrometry (EDS); (**d**) Appearance of a short surface fatigue crack on a replica film.

2.3.2. Four-Point Bending Fatigue Tests (R = 0.1)

In addition to the rotating bending fatigue tests at the stress ratio R = -1, four-point bending fatigue tests (inner span: 10 mm, outer span: 30 mm) were also conducted at R = 0.1 using the rectangular specimens shown in Figure 4c. A Vickers indentation was introduced on the specimen surface to initiate a crack. To eliminate the effects of the indentation, such as those of machining and residual stress, on the FCG behavior of short surface cracks, the introduced crack was extended to a length of approximately 200 µm by applying a cyclic load. A servo-hydraulic fatigue testing machine was used to perform the four-point bending fatigue test. The tests were conducted in air at room temperature with a stress frequency of 10 Hz. The FCG behavior of short surface cracks was measured using the replication technique.

3. Results

3.1. *S*-*N Curve* at R = -1

The *S*-*N* curve was investigated using the un-notched fatigue specimens. Figure 6 shows the *S*-*N* curve for the fine-grained WC-Co cemented carbide at a stress ratio of R = -1. The fatigue limit at R = -1 was estimated to be 1600 MPa. The solid line in the figure shows an approximate fitting curve to the experimental data. A fatigue test was conducted using one specimen for each stress amplitude, since a statistical analysis on the fatigue life is not the main purpose of this study.



Figure 6. S-N curve for the fine-grained WC-Co cemented carbide.

3.2. Fatigue Crack Growth (FCG) Behavior of Short Surface Cracks

The FCG behavior of short surface cracks in the WC-Co cemented carbide was studied using the notched specimens. Figure 7 shows the variation of the crack length 2*a* as a function of the number of cycles *N* under several conditions at R = -1. Figure 7a shows the result under a constant stress amplitude σ_a of 600 MPa. Figure 7b,c shows the 2*a* vs. *N* relations for constant K_{max} values of 6.0 and 7.5 MPam^{1/2}, respectively. The rate of FCG, da/dN, under the constant- σ_a test was obtained from Figure 7a as the slope of the approximate fitting curve to the experimental data, and K_{max} was calculated using Equation (1). The values of da/dN in the constant- K_{max} tests in Figure 7b,c were similarly evaluated as the slopes of the approximate straight lines fitted to the experimental data.

Figure 8a,b shows the relations between the rate of FCG, da/dN, and the K_{max} obtained at stress ratios of R = -1 and 0.1, respectively. In these figures, the experimental results are plotted for the constant- σ_a tests (open circles) and the constant- K_{max} tests (black and gray solid circles). Focusing on the effect of the stress ratio by comparing the results of Figure 8a, b, little difference due to the stress ratio can be seen in the da/dN vs. K_{max} relations.



Figure 7. Cont.



Figure 7. Variation of crack length 2a as a function of number of cycles *N* under several conditions at R = -1: (a) $\sigma_a = 620$ MPa; (b) $K_{max} = 6.0$ MPam^{1/2}; (c) $K_{max} = 7.5$ MPam^{1/2}.



Figure 8. Relations between da/dN and K_{max} for short surface fatigue cracks; (**a**) R = -1; (**b**) R = 0.1.

3.3. Observation of the Crack Growth Path

The crack growth path was investigated in detail for R = -1 and under the constant- K_{max} conditions (Figure 7b,c). Special attention was paid to the interaction between the fatigue crack and the microstructure (WC grains and the Co phase), particularly the effect of K_{max} on the FCG path. The observations were conducted in two ways. First, after interrupting the FCG test, the specimen surface was investigated in detail using the SEM. Second, after the failure of a specimen, its fracture surface was observed using the SEM. To clarify the relationship between the observations of the

fracture surface and specimen surface, the same locations of the fatigue specimen were selected for the two observations.

Figure 9 shows the observation results for K_{max} values of 6.0 MPam^{1/2} (Figure 9a) and 7.5 MPam^{1/2} (Figure 9b). These figures comprise (A) schematic illustrations of the crack growth path, (B) SEM images of the specimen surface, and (C) SEM images of the fracture surface. Labels (1–3) are attached to these figures to indicate the same locations in each figure. Some broken lines have been added to Figure 9a-B,b-B to facilitate the visualization of grain boundaries.

The observed crack growth paths on the specimen surfaces were classified into four types [17]: (1) within WC grains, (2) within the Co phase, (3) along the WC/WC boundaries, and (4) along the WC/Co boundaries. As can be seen from Figure 9a-A,a-B, at the low K_{max} value of 6.0 MPam^{1/2}, the crack mainly grew along the WC/WC boundaries (type (3)) among the above four types of crack growth paths. This finding can also be confirmed from the fracture surface in Figure 9a-C, where many missing portions of WC grains exist, as highlighted by the yellow arrows. This indicates that the crack grew along the WC/WC boundaries at the low value of K_{max} . In contrast, at the higher K_{max} value of 7.5 MPam^{1/2} (Figure 9b-A,B), many cleavage fractures of WC grains were observed compared with at the low value of K_{max} . This can also be confirmed from the fracture surfaces, where there are many cleavage faces, as highlighted by the yellow arrows in Figure 9b-C. Thus, the crack grew within WC grains more frequently at the higher value of K_{max} . In the fatigue crack growth process of cemented carbides, WC grains form bridging, which is usually formed behind the main crack tip. As indicated by the black arrow in Figure 9b-B, for example, the existence of bridging can be confirmed. It appears that there is another crack in addition to the main crack.

Figure 9 shows part of a crack growth path in detail. The observed crack growth paths on specimen surfaces with a large area were classified into four types, and the percentages of each type of crack growth path were evaluated at several K_{max} values. The results of the classification are shown in Figure 10, in which the percentages of the four types of crack growth path are shown as a function of K_{max} . The percentage of type (3) paths, i.e., along the WC/WC boundaries, decreased from 50 to 30% with an increase in K_{max} from 6.0 to 7.5 MPam^{1/2}, whereas the percentage of type (1) paths, i.e., within WC grains, increased from 20 to 40%.



(a) $K_{\text{max}} = 6.5 \text{ MPam}^{1/2} (\text{Low-}K_{\text{max}})$

1µm

Figure 9. Cont.



(**b**) $K_{\text{max}} = 7.5 \text{ MPam}^{1/2}$ (High- K_{max})

Figure 9. Observations of crack growth paths at R = -1. Tests were conducted at a constant value of K_{max} ; (a) $K_{\text{max}} = 6.0 \text{ MPam}^{1/2}$; (b) $K_{\text{max}} = 7.5 \text{ MPam}^{1/2}$; (A) Schematic illustrations of the crack growth path; (B) SEM images of the specimen surface; (C) SEM images of the fracture surface.



Figure 10. Relationship between percentage of each type of crack growth path and K_{max} , observed at R = -1.

Figure 11 shows SEM images of the fracture surface of an un-notched specimen that failed at σ_a of 1600 MPa. Figure 11a shows the fracture surface at the low K_{max} near the crack origin, and Figure 11b shows the fracture surface at the high K_{max} far from the crack origin. The K_{max} values for surface cracks with a half crack length *a*, were calculated using the following expression, where the crack length was measured from the crack origin:

$$K_{\max} = Y \sigma_a \sqrt{\pi a}.$$
 (2)

In this calculation, an internal crack slightly beneath the specimen surface was regarded as a surface crack with a semicircular shape. Therefore, a value of 0.73 was used for *Y*, as used in Equation (1). From the fracture surface observations, it was confirmed that the fatigue crack initiation site for the present material was an aggregate of several WC grains or a coarse WC grain slightly beneath the specimen surface [21].

There is a noticeable difference between the two images in Figure 11. On the fracture surface near the fracture origin in Figure 11a (at the low K_{max}), some missing portions of WC grains were also observed, as highlighted by the red arrows at the low K_{max} . In contrast, on the fracture surface far from the fracture origin shown in Figure 11b (at the high K_{max}), several cleavage fractures of the WC grains were observed, as highlighted by the blue arrows. The above observations of the fracture surfaces were in good agreement with the results shown in Figure 10.



(**b**) $K_{\text{max}} = 10.5 \text{ MPam}^{1/2}$

Figure 11. SEM images of the fracture surface on the un-notched specimen failed at σ_a of 1600 MPa. Red arrows show missing portions of WC grains. Blue arrows show cleavage fractures of WC grains; (a) Image taken near the crack origin (at the low K_{max}); (b) Image taken far from the crack origin (at the high K_{max}).

3.4. Effect of the Microstructure on the Rate of FCG

Figure 12 shows the variation of the rate of FCG, da/dN, for a short surface crack as a function of the microstructure observed along the FCG path. FCG tests were interrupted periodically to observe the crack. Then, the specimen was removed from the fatigue testing machine and placed in an SEM. The crack length was measured while the specimen was in the SEM. da/dN was measured in several

zones, indicated as (1–5) in the figure. da/dN in zone (3) (the cleavage fracture zone of WC grains) was higher than that in the other zones, indicating that the FCG resistance was lowest within WC grains. Crack tortuosity and bridging parts can be seen in zone (4), i.e., at the WC/WC or WC/Co boundaries. da/dN in this zone was much lower than in the other zones.



Figure 12. Changes in crack growth rate along the crack path at a constant K_{max} of 7.0 MPam^{1/2}.

4. Discussion

4.1. Difference in the FCG Behavior Between Short Surface Cracks and Long Through-Thickness Cracks

The FCG behavior of short surface cracks obtained in the present study was expected to be different from that of long through-thickness cracks. In the following, the FCG behavior of long through-thickness cracks [4] previously obtained using the same cemented carbide will be compared with that of short surface cracks. FCG tests on the long through-thickness cracks were conducted at a stress ratio of R = 0.1 and at a stress frequency of 10 Hz using CT (Compact Tension) specimens [4]. The result is shown in Figure 13. In this figure, the FCG data for long through-thickness cracks (stress ratio R = 0.1) obtained by other researchers [14] using a similar fine-grained cemented carbide to that used in the present study are also shown. The FCG data can be approximated by straight lines on the logarithmic graph, and the following Paris equations [24] were derived:

$$da/dN = CK_{\max}^m.$$
(3)

The power exponent *m* in the above equation was approximately 2 for short surface fatigue cracks but about 10 for long through-thickness fatigue cracks.

It is well known that residual stress on a specimen's surface strongly affects the crack growth behavior, even at small scales such as on thin films. For example, Ghidelli et al. [25] evaluated the elastic moduli on thin films considering the effect of residual stress. However, when the residual stress varies with time, as in fatigue crack growth, evaluation of the effects of residual stress often involves complicated stress analysis. Therefore, there have been very few studies on the FCG of cemented carbides considering the effects of residual stress. Thus, in this study, to study the FCG behavior,

the specimen surface was polished to a mirror-like finish to minimize the effects of compressive residual stress induced by machining. The obtained data was compared with that of other researchers, where the effects of residual stress were not taken into consideration, in contrast to in our study. This is the main reason for not considering the effects of residual stress on FCG in this study. More detailed research on the effect of residual stress on the FCG behavior of cemented carbide will be carried out in the future.

As can be seen from the figure, the short surface fatigue crack has a longer region where the fatigue crack grows stably than the long through-thickness fatigue crack. Thus, the FCG behaviors of short surface and long through-thickness fatigue cracks are clearly different.

The K_{max} value (hereinafter referred to as K_{fc}), at which the crack begins to grow unstably ($da/dN > 10^{-6}$ m/cycles) was determined to be 10 MPam^{1/2} for the short surface fatigue crack, compared with 7.5 MPam^{1/2} for the long through-thickness fatigue crack. It was found that the long through-thickness crack grows in a more brittle and unstable manner than the short surface fatigue crack.

In Figure 13, the K_{max} value at the fatigue crack growth rate da/dN of 10^{-6} m/cycles is defined as K_{fc} and that at a da/dN of 10^{-9} m/cycles is defined as K_{th} .

Figure 14 shows the change in the ratio of $K_{\text{th}}/K_{\text{fc}}$ as a function of the average thickness of the Co phase, λ_{Co} . In this figure, the experimental results obtained by Fry et al. [12] and by Llanes et al. [14] are also plotted. In their studies, similar cemented carbides to that in the present study were used. Table 3 shows the WC grain sizes and Co contents of the cemented carbides whose FCG data were plotted in Figure 14.

As shown in Figure 14, the tendency of fatigue susceptibility (K_{th}/K_{fc}) in the crack growth behavior is not greatly affected by the difference in the WC-Co components of the materials. In other words, the difference between a long through-thickness crack and a short surface crack can be clearly seen even if the WC-Co components of the materials are different. The values of K_{th}/K_{fc} for long through-thickness fatigue cracks at R = 0.1 (black solid marks) were clearly larger than those of short surface fatigue cracks at R = -1 and 0.1 (open or gray circles). A clear difference was observed between the both types of crack. It is pointed out [14] that the mode of static fracture increases as the value of K_{th}/K_{fc} increases, and the fatigue behavior is strengthened as the value of K_{th}/K_{fc} decreases. Therefore, the results in Figures 13 and 14 show that the FCG behavior of short surface cracks is more susceptible to repeated stress than that of long through-thickness cracks.

- \bigcirc R = -1 (Short; fine-grained 13%Co alloy)
 - \bigcirc *R* = 0.1 (Short; fine-grained 13%Co alloy)
 - \triangle R = 0.1 (Long; fine-grained 13%Co alloy) [4]
 - \blacksquare R = 0.1 (Long; suitable alloy for fine-grained 10%Co alloy) [14]
 - \square *R* = 0.1 (Long; suitable alloy for fine-grained 16%Co alloy) [14]



Figure 13. FCG rate for several fine grained alloys as a function of K_{max} values.



Figure 14. Fatigue sensitivity and microstructure relationship for fine-grained WC-Co cemented carbides.

Table 3. Microstructural properties and fatigue-testing conditions for several fine-grained cemented carbides.

5	Symbols Show in Figure <mark>14</mark>	ⁿ Co% (wt %)	d _{WC} (μm)	λ_{Co} (µm)	K _{Ic} (MPam ^{1/2})	Crack Length	R (-)	Reference
	• a	13	0.45	0.09	12.4	Long	0.1	Mikado et al. [4]
	• b	6	1.0	0.093	8.2	Long	0.1	Fry et al. [12]
	• c	6	0.39	0.15	7.5	Long	0.1	Llanes et al. [14]
	• d	10	1.1	0.216	9.1	Long	0.1	Fry et al. [12]
	• e	10	0.50	0.25	9.2	Long	0.1	Llanes et al. [14]
	O f	13	0.45	0.09	12.4	Short	$^{-1}$	This work
	O g	18	0.45	0.11	13.7	Short	$^{-1}$	Mikado et al. [24]
	Q h	13	0.45	0.09	12.4	short	0.1	This work

In the case of short surface fatigue cracks, the crack size is closer to the microstructure size of the material than for long through-thickness fatigue cracks. Thus, it is suspected that their FCG property is affected by the microstructure. As can be seen from Figure 9, the short surface fatigue cracks grow along the WC/WC grain interface in a zigzag manner while bypassing WC grains. Furthermore, as shown in zone (4) of Figure 12, a bridging effect is expected to occur, in which unbroken hard particles connect the upper and lower faces of the crack. Therefore, the above zigzag growth and bridging effect are more likely produced in the case of short surface fatigue cracks than in the case of long through-thickness fatigue cracks, which are followed by a decrease in the FCG driving force at the crack tip. As a result, it can be inferred that short surface fatigue cracks will grow more stably than long through-thickness fatigue cracks.

In a metal, it is well known that a short surface fatigue crack has a higher FCG rate than a long through-thickness fatigue crack at a low stress intensity factor range ΔK . A short surface fatigue crack will also grow under the threshold level of the stress intensity factor range ΔK_{th} for a long through-thickness fatigue crack. In the present study, FCG data on the cemented carbide have not been systematically obtained in the low fatigue crack growth rate range, thus the difference between the FCG behavior of short surface cracks and long through-thickness cracks at the low values of K_{max} has not yet been clarified. Further study on this issue is required.

4.2. Effect of the Maximum Stress Intensity Factor, K_{max}, on the Crack Growth Path

To consider the effect of K_{max} on the crack growth path, the size of the plastic zone r_{p} at the crack tip was calculated using

$$r_{\rm p} = \frac{1}{2\pi} \left(\frac{K_{\rm max}}{\sigma_{\rm y}} \right)^2,\tag{4}$$

where σ_y is the yield strength of the material, which is unknown at present. Thus, the bending strength indicated in Table 2 was used instead of the yield strength. The calculated values of r_p for each value of K_{max} are shown in Table 4. The calculation of the plastic zone size at the crack tip used in this paper is applicable only to a single-phase material. However, this calculation method was applied to easily obtain the plastic zone size at the crack tip in the material. Similar results were also obtained in the two-dimensional FEM (Finite Element Method) analysis reflecting the actual microstructure of the cemented carbide.

Table 4. Plastic zone sizes at the crack tip calculated for several values of K_{max} .

$K_{\rm max}$ (MPam ^{1/2})	<i>r</i> p (μm)		
6.0	0.34		
6.5	0.40		
7.0	0.46		
7.5	0.53		

The size of the plastic zone r_p at the crack tip increases with K_{max} as expected. At the low K_{max} of 6.0 MPam^{1/2}, r_p of 0.34 µm is smaller than the average WC grain size (0.45 µm) of the present alloy. In this case, it is difficult for a crack to grow within WC grains, i.e., for transgranular cracking to occur. Instead, the crack is expected to grow preferentially along the weak WC/WC interfaces. On the other hand, at the high K_{max} of 7.5 MPam^{1/2}, r_p of 0.53 µm is larger than the average WC grain size. In this case, the crack grows within WC grains and tends to be straighter than that for a low value of K_{max} . It grows more rapidly because of the higher crack driving force and the fewer bridging parts than for the low value of K_{max} . This mechanism is schematically illustrated in Figure 15.



Figure 15. Schematic illustration of the effect of K_{max} on the FCG path: (**a**) at low K_{max} (WC/WC interface, bridging parts); (**b**) at high K_{max} (cleavage fracture of WC grains).

5. Conclusions

Rotating and four-point bending fatigue tests were conducted on a fine-grained WC-Co cemented carbide to investigate the FCG behavior of short surface fatigue cracks in detail. Then, the differences in the FCG behavior of short surface fatigue cracks and long through-thickness fatigue cracks were studied. The following conclusions were obtained.

- 1. There was little effect of the stress ratio on the $da/dN-K_{max}$ relation for short surface fatigue cracks.
- 2. Short surface fatigue cracks have a longer region where fatigue crack growth is stable than long through-thickness fatigue cracks. In other words, long through-thickness fatigue cracks grow in a more brittle and unstable manner than short surface fatigue cracks. The FCG behavior of the short surface cracks is more susceptible to repeated stress than that of long through-thickness cracks. Thus, the FCG behaviors of both types of crack are clearly different.

- 3. The crack growth paths were classified into four types: (1) within WC grains, (2) within the Co phase, (3) along WC/WC boundaries, and (4) along WC/Co boundaries. The percentage of crack growth paths along WC/WC boundaries decreased from 50 to 30% when K_{max} was increased from 6.0 MPam^{1/2} to 7.5 MPam^{1/2}, while the percentage of transgranular cracks within the WC grains increased from 20% to 40%.
- 4. At low values of K_{max}, it is not easy for a crack to grow within WC grains (transgranular cracking). Instead, fatigue crack growth occurs discontinuously along weak WC/WC interfaces or along WC/Co boundaries. On the other hand, at high values of K_{max}, cleavage fractures of WC grains are more common and the area in which cleavage fractures occur is wider than at low values of K_{max}. Moreover, cracks tend to be straighter than those for a low K_{max}. Thus, cracks grow more rapidly for a high K_{max} because of the higher crack driving force and the fewer bridging parts.

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