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The influence of microstructure on the mechanical properties of polycrystalline diamond: a literature review*

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ABSTRACT
Polycrystalline diamond (PCD) is an extremely high-performance cutting tool material used in the machining of rock, high-strength, non-ferrous metal alloys and carbon-fibre-reinforced composites. It is favoured for its exceptional hardness and wear resistance which results in at least an order of magnitude improvement in performance over previous technologies in almost all metrics. However, PCD suffers from unpredictable brittle fracture and degradation at high temperature during service which limits its capabilities in cutting applications. The literature on the link between its microstructure and its mechanical properties, including strength, toughness and flaw size distribution as measured by pseudo-static tests, is investigated. The conclusions of the seminal paper on this topic are re-examined in the light of modern ceramics research and an alternative explanation is put forth for the strength–grain size relationship published in this paper. All known literature values for strength and toughness vs. grain size and binder content are collated showing no overall trend in strength with binder content but moderate trends in all other combinations. The common claim of weak grain boundaries is brought into question in the light of the lack of any evidence of this fracture mode being evident in pseudo-static tests. The industrial literature on wear testing and failure modes of PCD in service and service-like tests is examined to bridge the gap between pseudo-static and dynamic, application-based experiments. Six main failure modes are recorded and summarised with intergranular fracture being the most conspicuously absent from the pseudo-static tests. It is suggested that the temperature generated by friction in dynamic tests causes the weakening of grain boundaries, resulting in a transition from transgranular to intergranular fracture and a call for further research in this area is made.

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KEYWORDS
Polycrystalline diamond; PCD; microstructure; mechanical properties; fracture toughness; strength; flaw size distribution; thermal degradation

Introduction

Brief history of diamond synthesis
Almost as soon as diamond was discovered to be an allotrope of carbon in 1797 by Smithson Tennant [1], efforts were made to produce it synthetically. Many claimed success but a distinct lack of evidence and reproducibility suggests that these early reports were fallacious. Simultaneously the most famous and tragic of which was that of Henri Moissan. He attempted to use the solubility of carbon in iron combined with a rapid quench from white heat to induce the pressure required to transform graphite into diamond [2]. Although he claimed success, testimony from his widow stated that one of his assistants had introduced natural diamond flakes into his samples ‘to please the old man’ and cut short his tediously long experiments [3]. Later cathodoluminescence work confirmed that the diamonds were indeed natural in origin [4].

It was not until 1955 [5] (with further details given in 1959 [6]) that the first credible report of diamond being synthesised was published by researchers at General Electric. They discovered that a molten metal could be used to lower the temperatures and pressures needed to synthesise diamond making it feasible given the technology of the time but did not understand the mechanism of growth. The role of the molten metal was strongly contested until a study in 1981 concluded that it must play a role as both a solvent for graphitic carbon and a catalyst for the re-precipitation of diamondiferous carbon [7], lowering the activation energy barrier of its formation.

Polycrystalline diamond

History
The first instance of a sintered, polycrystalline diamond (PCD) compact was published in 1970, 12 years after it was predicted that a synthetic form of natural PCD, carbonado, would be a highly desirable material to be able to make [8] by the same author. The following year, it was found that using cobalt as a binder could reduce the temperature and pressure required for successful sintering by up to 24 and...
27%, respectively [9]. This technological innovation allowed the mass production of PCD to be realised and still accounts for the way in which most PCD is made today.

In 1973, General Electric patented the use of a carbide-supported PCD drill bit for use in oil and gas drilling [10]. From then onwards, PCD cutters made inroads into various material removal industries such as the machining of wood-based products [11], non-ferrous metals [12] and most recently, composites [13]. In 2010, PCD cutters accounted for 65% of the footage drilled by the oil and gas industry [14], projected to rise to over 80% by 2016 [15]. Such market domination has been due in large part to the cost savings made possible by PCD’s remarkable properties. PCD cutters provide up to two orders of magnitude improvement in tool life over other industry standard-cutting materials such as cemented tungsten carbide depending on the specific application [16–18]. This allows far less time and money to be expended on replacing worn tools; something that is particularly costly in drilling applications, leading to vastly reduced operating costs and hence increased profits. Outstanding hardness and abrasion resistance are the two main materials properties of PCD that have allowed such huge performance increases compared to other materials. But, as ever, with high hardness comes low toughness. Although PCD is far less brittle than single crystal diamond, its resistance to impact and shock loading are two of the main issues that must be overcome in improving PCD as a cutting material.

**Synthesis of PCD**

PCD synthesis begins with diamond powder; this can consist of a single carefully controlled grain size, a wide distribution of grain sizes or multiple single sizes mixed together. The powder is compacted on the top of a cemented tungsten carbide substrate which may have a textured top surface to mitigate the residual stresses that develop at the diamond–substrate boundary. This is encapsulated by a metal canister and inserted into a capsule that in turn goes into a high pressure press. There are two common designs of press: belt and cubic. Cubic presses apply pressure along each of the three Cartesian axes, whereas belt presses apply pressure from above and below and constraint circumferentially. Presses typically apply 1400–1800°C and 4–10 GPa of pressure. The capsule is designed so that a constituent of it becomes molten at high temperature, distributing the applied pressure hydrostatically.

The four major processes that occur during the sintering of PCD are outlined in Figure 1. The resultant product is a layer of two-phase ceramic predominantly comprised of micron-sized diamond grains sintered together with a high degree of diamond–diamond bonding atop a thick tungsten carbide/cobalt substrate.

The diamond grains retain roughly the same morphology post-sintering and the cobalt binder occupies the spaces between the grains. The micro-structure is a complex combination of the original diamond grains, newly precipitated diamond [19], metal binder and various intermetallic phases [22]. In addition, there are regions of plastically deformed material [20] and quite substantial microscopic and macroscopic residual stresses [23].

Leaching of PCD

Although cobalt is very useful in acting as a catalyst for diamond synthesis, it unfortunately also contributes to the thermal degradation of PCD under the high-temperature conditions of drilling [24]. A patent was filed in 1978 describing a method of removal of the cobalt binder phase to increase PCD’s thermal stability, called leaching [25]. Leaching involves heating PCD in a mixture of concentrated acid for long periods of time to dissolve the cobalt. While this does increase the thermal stability, the mechanical properties are degraded [26] and it is a time-consuming and dangerous process. Electrolysis has also been used to leach PCD [27] with good results. Few studies have been done on the effect leaching has on the mechanical properties of diamond and these will be discussed in later sections.

Traditional mechanical testing of PCD

Techniques, such as notched and unnotched beam flexure, ball-on-three ball flexure and diametral compression, are discussed in this section. The term ‘traditional’ is taken to mean mechanical deformation of macroscopic samples. Almost all of the work on diamond’s mechanical properties cite the work of Lammer [28] from 1988. A recent review paper entitled ‘The mechanical and strength properties of diamond’ by Field quotes it as if it is the state of the art [29]. Although Lammer’s work made significant improvements over the early work carried out by Gigl [30] and Roberts [31], there are several points to be addressed regarding the conclusions of the paper.

The seminal paper on the mechanical properties of PCD

The first work to examine the mechanical properties of PCD in great depth was written by Lammer in 1988. This paper presented a big step forward in the understanding of this research area and still influences work undertaken today. However, some of its conclusions warrant a re-examination in the light of modern understanding of ceramic materials and given its prominence in the field, the entirety of the present section has been dedicated to doing just that.

Figure 2 illustrates the relationship between transverse rupture stress and the inverse of the square root of nominal grain size deduced by Lammer for various
PCD grades. It shows two distinct regions in behaviour. The bimodal nature of the curve is attributed to an explanation by Rice [32] based on work on other ceramics showing that the inherent critical flaw size $c$ is independent of the grain size $d$. The argument continues that for fine-grained materials, where the inherent flaws may span numerous grains, the strength is controlled by the polycrystalline fracture toughness. For coarse grains, where the critical flaw is within a single grain, it is controlled by something akin to a single-crystal fracture toughness. The crossover point between these two regimes is expected to be where grain size, $d$, is equal to flaw size, $c$. From Table 1, $c/d = 1$ occurs between 12–30 $\mu$m which agrees with the approximate crossover point between 'region 1' and 'region 2' in Figure 2 of 30 $\mu$m.

However, the grain size used in this analysis is the average diameter of particles in the original powder prior to sintering. During the high-pressure sintering process, particles are crushed due to large stresses at point contacts [21], widening the particle size.
distribution and pushing it to lower diameters. This process is much more severe for large particles as the number of contact points is fewer causing higher local stresses. As particle size decreases, the comminution limit is eventually reached meaning little or no crushing occurs. Therefore, the assertion that the crossover point occurs where \( c = d \) is likely untrue as ‘nominal grain size’ largely, but not linearly, overestimates the true grain size.

Additionally, two distinct fracture behaviours would be expected to be seen. The ‘single crystal fracture toughness’ for large grain sizes and the ‘polycrystalline fracture toughness’ for small-grain sizes. It is also implied that large-grained materials should exhibit transgranular fracture while small-grained materials show intergranular fracture. As can be seen from Figure 3, no such clear distinction is evident and although fracture in large-grained (\( \geq 30 \mu m \)) material is described as being transgranular, there is no mention of the fracture mode of fine-grained materials and no corroborating SEM images. The explanation given for the observed trend is derived from Rice’s work stating that ‘thermal and elastic anisotropy between diamond and binder phases could result in high interfacial stresses that allow micro-cracks to propagate along the diamond-bonding boundaries’. While high-residual stresses have been observed due to the thermal expansion mismatch of diamond and cobalt [33], fracture occurs predominantly transgranularly in PCD [34] so micro-cracks along the grain boundaries, if they do exist, have minimal effect on the fracture behaviour. He then goes on to say that other factors such as ‘impurities in the matrix material, the amount and size distribution of the secondary phase, and the degree of plastic deformation and fragmentation of diamond grains during synthesis’ could affect the fracture toughness. Some of these postulations are almost certainly true, but further work is required to show which hold merit.

As it has been established that the flaw size is not independent of the grain size, Lammer’s data can be reinterpreted. It is commonly found in ceramics that for large-grained materials, the individual grains act as critical flaws and so the fracture behaviour follows Irwin’s modification of the Griffith equation

\[
\sigma_c = \frac{K_{IC}}{\sqrt{\pi c}}
\]

‘Region 1’ in Figure 2 is a straight line that passes approximately through the origin on a graph of strength \( \sigma_c \) against inverse, root grain size \( d^{-1/2} \), suggesting that this typical behaviour is indeed followed. Also given the roughly constant value of \( c/d \) for grain sizes above 30 \( \mu m \) in Table 1 it can safely be assumed that the flaw size is proportional to the grain size in this region.

By comparing Equation (1) to the gradient of ‘region 1’, the implied fracture toughness of the material can be calculated from Equation (2) by substituting \( Y = 2/\pi \) and \( c = d/2 \), assuming that the flaws are semi-elliptical and surface breaking

\[
K_{IC}^{imp} = 1.12 \cdot m \sqrt{\frac{2}{\pi}}
\]

where \( m \) is the gradient of ‘region 1’.

The result of this calculation gives \( K_{IC}^{imp} = 6.67 \) MPa\( \sqrt{m} \). Figure 4 shows this result along with the measured fracture toughness of the materials in ‘region 1’ (materials C-G, values taken from Figure 3). Lammer’s measurements follow the work of Devin et al. [35], using laser cut notches in the centre of diametrical compression samples. If the notch created by this technique has a much greater radius that the inherent flaw size of the material, it is considered blunt and will overestimate the toughness [36]. A treatment by Fett and Munz [37] calculates the true fracture toughness
from the apparent value by assuming a small, incipient, sharp crack protrudes from the blunt notch using the following equation:

\[ K_{\text{true}} = K_{\text{app}} \tanh \left( \frac{2Y}{\rho} \right) \]  

where \( Y \) is the geometrical shape factor, \( \delta a \) is the incipient crack length and \( \rho \) is the notch-root radius. Given that Equation (3) reduces the apparent fracture toughness by a greater amount for smaller incipient crack sizes (and hence smaller grain sizes), the calculated value of \( K_{\text{imp}} \) in Figure 4 and the measured toughness values are in reasonable agreement and a single-valued \( K_{\text{Ic}} \) in ‘region 1’ is plausible. Unfortunately, the notch-root radius was not given and so the actual adjustment cannot be performed.

If ‘region 2’ merely had a different, single-valued polycrystalline fracture toughness, it would still be expected that the data would form a straight line through the origin, only with a different gradient. This is evidently not the case. Instead, a much weaker dependence on grain size is seen, suggesting that another factor is controlling the strength.

Data from other authors measuring the strength of PCD are tabulated in Table 2. Belnap and Griffo [41] compared the strength of PCD monoliths and PCD composites with additional tungsten carbide/cobalt admixed. They found that decreasing the diamond content of the composite resulted in a decrease in the flexural strength. The two different grain sizes they used resulted in similar flexural strengths within the bounds of the error. Along with showing that there is a negative strain rate and positive temperature dependence on strength, Petrovic et al. [38] found that a larger grain size resulted in a lower strength material. This was ascribed to larger flaws found in the coarser-grained material; they noted that the typical flaw size found in the surface of the specimen was proportional to the grain size. Despite reporting the strength of one grade of PCD, Morrell et al. [40] do not provide any additional information on which grade this was. Finally, McNamara et al. [39] found that a small decrease in the cobalt content of the PCD resulted in a small decrease in the strength.

The data in Table 2 are also shown graphically for grain size and binder content in Figures 5 and 6. From Figure 5, a reasonably clear trend can be seen that decreasing the grain size increases the strength. This agrees with the general trend shown by Lammer but two distinct regions are not obvious in these data. The correlation between flaw size and grain size reported by Petrovic et al. disagrees with Lammer’s suggestion that flaw size is independent of grain size; however, further results are needed to verify the trend and an investigation into the flaw size distribution for various grades of PCD should be carried out.

In Figure 6, a less obvious trend is seen between strength and binder content, although there may be a general positive correlation, the large scatter in the data adds significant uncertainty. As it is difficult to vary the binder content independently of grain size, understanding the role of the binder phase on the mechanical properties from literature data is problematic.

**Table 2.** Strength data for various grades of PCD collated from numerous authors.

<table>
<thead>
<tr>
<th>Author</th>
<th>Technique</th>
<th>Strength (MPa)</th>
<th>Weibull modulus</th>
<th>Grain size (μm)</th>
<th>Binder content (wt-%)</th>
</tr>
</thead>
<tbody>
<tr>
<td>Petrovic et al.</td>
<td>3PB</td>
<td>750 ± 75</td>
<td>–</td>
<td>6</td>
<td>23</td>
</tr>
<tr>
<td></td>
<td>650 ± 75</td>
<td>–</td>
<td>30</td>
<td></td>
<td></td>
</tr>
<tr>
<td>McNamara et al.</td>
<td>-</td>
<td>1075 ± 82</td>
<td>13.4</td>
<td>–</td>
<td>8.5</td>
</tr>
<tr>
<td></td>
<td>1034 ± 129</td>
<td>8.3</td>
<td>–</td>
<td></td>
<td></td>
</tr>
<tr>
<td>Morrell et al.</td>
<td>B3B</td>
<td>1453 ± 29</td>
<td>14.5</td>
<td>–</td>
<td></td>
</tr>
<tr>
<td>Belnap and Griffo</td>
<td>3PB</td>
<td>1815 ± 100</td>
<td>16.0</td>
<td>3.5</td>
<td>23.5</td>
</tr>
<tr>
<td></td>
<td></td>
<td>1714 ± 95</td>
<td>21.9</td>
<td>5</td>
<td>20</td>
</tr>
<tr>
<td>Lammer [28]</td>
<td>DC</td>
<td>1550 ± 185</td>
<td>3.4</td>
<td>2</td>
<td>13</td>
</tr>
<tr>
<td></td>
<td>1256 ± 175</td>
<td>4.3</td>
<td>12</td>
<td></td>
<td>11</td>
</tr>
<tr>
<td></td>
<td>1180 ± 175</td>
<td>4.4</td>
<td>30</td>
<td></td>
<td>11</td>
</tr>
<tr>
<td></td>
<td>444 ± 150</td>
<td>3.4</td>
<td>125</td>
<td></td>
<td>12</td>
</tr>
<tr>
<td>Marro et al. [42]</td>
<td>B3B</td>
<td>1610 ± 230</td>
<td>–</td>
<td>4</td>
<td>10</td>
</tr>
<tr>
<td></td>
<td>1120 ± 160</td>
<td>–</td>
<td>10</td>
<td></td>
<td>7.5</td>
</tr>
<tr>
<td></td>
<td>1000 ± 200</td>
<td>–</td>
<td>20</td>
<td></td>
<td>6</td>
</tr>
</tbody>
</table>

The following terminology is used for the technique column: 3PB – 3 point bending, B3B – ball-on-three-ball, DC – diametral compression. The error shown in the strength column is the standard deviation except for Lammer where it is the range in the measured results.
Fracture toughness

Miess and Rai [43] investigated the fracture toughness of a wider range of grain sizes using the same method as Lammer. Their results show a plateau in fracture toughness rather than the drop-off found by Lammer. Along with citing Rice’s work, they suggest that as grain boundaries in ceramics are typically weaker than the grains, increasing the area of grain boundary by decreasing the average grain size increases the amount of weak material and hence lowers the fracture toughness. This argument suffers from the same flaw as that of Lammer – no fracture paths are shown in their work and later work has shown that fracture in PCD occurs predominantly transgranularly.

Double-torsion tests were carried out on various grades of PCD by Lin et al. [44] to measure the fracture toughness and investigate sub-critical crack growth. They measured unusually high values of fracture toughness, in the region of 13 MPa√m. A review on the technique [45] notes that the use of a guiding groove, as was done in their work, is now recognised to be undesirable as the effects of the stress concentration it causes are poorly understood [46]. It is also noted that crack deflection is a sign of a poorly aligned or balanced loading system. It is unsurprising that Lin et al. observed crack deflection as with diamond’s exceptionally high stiffness, even the smallest misalignment will result in large-stress gradients. If the alignment of the specimen was slightly off, the stress intensity at the crack tip could be much lower than expected, resulting in the unusually high value of fracture toughness observed. However, despite its potential disadvantages, the results of this work are particularly interesting as a natural crack is grown in the sample rather than the artificial notches used in other tests.

Lin et al.’s work shows several micrographs of transgranular fracture, Figures 7 and 8, further suggesting that PCD does not preferentially fracture

Figure 5. Graphical representation of the strength vs. grain size data from Table 2. Studies were omitted if they did not have sufficient data on grain size. A general positive trend can be seen.

Figure 6. Graphical representation of the strength vs. binder content data from Table 2. Studies were omitted if they did not have sufficient data on binder content. The scatter in the data is large and no obvious overall trend can be seen.

Figure 7. Backscattered electron SEM image of a mostly straight crack path showing transgranular fracture. Its microstructure is described as ‘coarse diamond grains (20–50 μm) embedded in a fine-grained diamond matrix, containing little cobalt’. From Lin et al. [44].

Figure 8. Higher magnification backscattered electron SEM image of a crack running through a larger diamond grain. The authors describe this as indicative of crack deflection; however, it appears the degree of deflection is relatively small. From Lin et al. [44].
along the grain boundaries. The authors state that crack deflection appears to be a significant toughening mechanism in PCD, however, by examining the micrographs provided it can be seen that this is a minimal effect. It is also explained that in three of the samples tested, the crack advanced in discrete jumps with a stress relaxation occurring after each jump. The authors suggest that this may be due to the presence of the cobalt phase, presumably the plasticity therein, but due to only sporadic description of the microstructures of the samples tested, it is impossible to form any kind of correlation.

While others have measured the fracture toughness of PCD using single-edged v-notched beam (SEVNB) bending [40, 47], the most recent and in-depth study of this type is by McNamara et al. [48]. However, the analysis used in this paper, the theory of critical distances, was originally developed for use with metal components [49]. The 'critical distance' $r_c$ is initially taken to be the nominal grain size - a parameter which has already been seen to be inappropriate when discussing sintered PCD micro-structures. Upon realising the difficulty in defining it meaningfully, the authors surmise that it is merely a fitting parameter. Figure 9 shows that a small change in $r_c$ results in a large change in the calculated fracture toughness while still producing an acceptable fit to the experimental data. The authors note that this is not as catastrophic an issue for fine-grained micro-structures as it is for coarse grained as the degree of particle crushing is not as great, simplifying the choice of $r_c$.

McNamara et al. also studied the effect that ‘residual metal removal depth’ (leaching depth) had on the fracture toughness of PCD. Figure 10 shows that for coarse-grained material, the toughness remains constant until a certain depth and that fine-grained material’s toughness begins to degrade linearly as soon as leaching begins. It is unclear exactly why this trend exists, but the authors suggest that fine-grained PCD may be more sensitive to leaching as it has a higher volume fraction of cobalt and hence there will be more pores on the surface to act as flaws post-leaching.

The collated data on fracture toughness are shown in Table 3. Figure 11 shows that when aggregated, there is no discernible trend in fracture toughness against grain size. However, within each individual study, a similar trend is apparent – that increasing the grain size increases the fracture toughness but with increasingly diminishing effect. The main conclusion that can be drawn from this is that the changes in fracture toughness as the grain size varies are vastly outweighed by another effect that varies between studies. One possible issue is that three different testing methodologies are used within this data set. Even within a single-testing method such as SEVNB bending, different studies can be rendered incomparable if the notch roots are of different radii due to its strong effect on the apparent toughness [50]. It is highly likely that a similar effect will be present between different testing methodologies.

Figure 12, on the other hand appears to tell a completely different story. Of the four plots of collated data from the literature, fracture toughness against binder content appears to be the strongest. There seems to be a strong negative relationship between the toughness and the amount of binder in the PCD which is surprising given that others have suggested [34] that the binder phase contributes a toughening effect. However, the interaction between cracks in

---

**Figure 9.** Fitting of various values of critical distance $r_c$ to experimental data by McNamara et al. [48]. It can be seen that varying $r_c$ from 1 to 30 μm changes the calculated value of fracture toughness $K_{IC}$ dramatically while maintaining an acceptable fit to the experimental data.

**Figure 10.** Behaviour of the fracture toughness of various grades of PCD as the leaching depth (RMRD) is increased. The number in each label refers to the original diamond particle size. Materials A, B and C contained 8.5, 7.5 and 6.5% binder phase, respectively. It is unclear whether this is volume or weight percentage. From McNamara et al. [48].
PCD, the binder phase and the overall toughness of the material are likely much more complicated than this graph appears to show. For example, the binder content does not vary independently of grain size and the methodology effect may confound the results further. One must be wary of drawing any concrete conclusions from the trend shown in this graph. Although there appears to be a low amount of scatter, the results of each study are tightly grouped in binder content space rather than each showing a similar trend. A more carefully controlled investigation into the effect the binder has on fracture toughness is therefore required.

Flaw size distribution

A piece of information conspicuous by its absence from the literature on PCD is the flaw size distribution. Despite being briefly discussed by Lammer and Petrovic et al., no substantive effort to understand its influence has been published. As indicated in previous sections, it is likely that this information would help to explain the trends seen in the strength of PCD. One such technique that can probe the surface flaw size distribution in diamond is Hertzian indentation [51].

The first example of Hertzian indentation of diamond was by Howes and Tolansky [52] who were

Table 3. Fracture toughness data for various grades of PCD collated from numerous authors.

<table>
<thead>
<tr>
<th>Author</th>
<th>Technique</th>
<th>Fracture toughness (MPa√m)</th>
<th>Grain size (μm)</th>
<th>Binder content (wt-%)</th>
</tr>
</thead>
<tbody>
<tr>
<td>Petrovic et al. [38]</td>
<td>SDC</td>
<td>3.4 ± 0.34</td>
<td>2</td>
<td>30</td>
</tr>
<tr>
<td></td>
<td></td>
<td>4.4 ± 0.44</td>
<td>4</td>
<td></td>
</tr>
<tr>
<td></td>
<td></td>
<td>2.3 ± 0.23</td>
<td>6</td>
<td>27</td>
</tr>
<tr>
<td></td>
<td></td>
<td>4.5 ± 0.45</td>
<td>16</td>
<td></td>
</tr>
<tr>
<td></td>
<td></td>
<td>5.0 ± 0.50</td>
<td>28</td>
<td>18</td>
</tr>
<tr>
<td></td>
<td></td>
<td>6.0 ± 0.60</td>
<td>29</td>
<td>18</td>
</tr>
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<td></td>
<td></td>
<td>5.0 ± 0.50</td>
<td>30</td>
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<td></td>
<td></td>
<td>5.8 ± 0.58</td>
<td>45</td>
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<td></td>
<td></td>
<td>5.6 ± 0.56</td>
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<td>6.4 ± 0.64</td>
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<td></td>
<td></td>
<td>6.0 ± 0.60</td>
<td>120</td>
<td></td>
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<td>McNamara et al. [39]</td>
<td>3PB</td>
<td>9.55</td>
<td>10</td>
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<tr>
<td></td>
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<td></td>
<td></td>
<td></td>
<td>8.52</td>
<td>10</td>
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<td>Lin et al. [44]</td>
<td>DT</td>
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<td></td>
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<td>13.43 ± 0.66</td>
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<td></td>
<td></td>
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<td>SDC</td>
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<td></td>
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<td>8.81 ± 0.46</td>
<td>12</td>
<td>11</td>
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<td>8.89 ± 0.37</td>
<td>30</td>
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<td>7.49 ± 0.78</td>
<td>125</td>
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<td>Devin et al. [35]</td>
<td>SDC</td>
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The following terminology is used for the technique column: SDC - slotted diametral compression, 3PB – 3 point bending, DT – double torsion. The error shown in the strength column is the standard deviation in the results.
attempting to artificially induce the small, polygonal crack figures that Tolansky and Halperin [53] had seen on the octahedral faces of many natural diamonds. Lawn and Komatsu [54] then published a paper explaining that the mechanism behind the deformation that occurred when one of these so-called ‘pressure cracks’ was formed was completely elastic in nature. Field and Freeman [55] later used Hertzian indentation to measure the fracture toughness of single-crystal diamond. By plotting the square of the applied load vs. the cube of the cone-crack-base radius, something that is considerably easier to measure in transparent materials, the gradient can be used to calculate the fracture energy [56] and hence the fracture toughness. This was calculated as 3.4 MPa√m.

Despite this early work into single crystal and natural diamonds, it was not until 2015 that the technique was applied to PCD. Marro et al. [34] performed Hertzian indentation on various grades of PCD under both monotonic and cyclic loading conditions. Although the technique of acoustic emission is reasonably well documented in the literature on Hertzian indentation, it was not used to detect fracture in this study. The authors instead loaded samples of PCD with a spherical indenter up to predetermined loads and then examined them in an optical microscope to determine whether or not fracture had occurred. Although the data they collected was distinctly qualitative in nature, they did find that a fine-grained micro-structure was more resistant to the initiation of a ring crack as compared to coarse-grained specimens. The authors present micrographs that show local toughening by means of a reduced crack-opening displacement as the crack passes through a binder pool (Figure 13). Initially, this seems to disagree with the trend shown in Figure 12 as an increased binder percentage appears to result in a less tough material which could be taken to indicate that the binder phase is more prone to fracture than the diamond. However, it would appear that interactions are more complicated than that simplistic view. The material shown in Figure 13 comprises 10 wt-% binder which from Figure 12 suggests it should be a highly tough material. In light of this, it seems possible that the binder phase can still provide a toughening effect despite an increased percentage of it not resulting in an overall tougher material.

Figure 14 shows SEM images of crack paths in a coarser-grained micro-structure. The crack bifurcates as it passes through a large-diamond grain. From the information given and the background literature, it is quite unclear why this might happen.

The second paper on this topic published by Marro et al. deals with the damage tolerance of PCD materials...
roughly constant dramatically but that of the coarse-grained remains why the strength of the fine-grained material drops in the coarse-grained PCD. This would then explain material but of a similar size to those already present be much larger than the flaws in the fine-grained Hertzian indentation, the ring crack introduced could material can be explained by larger initial flaws. After indentation by measuring the residual strength by means of ball-on-three-ball flexure [42]. This is a very relevant topic for PCD research as one can imagine Hertzian-like fractures being caused by loaded contact with rock asperities which could then lead to catastrophic failure of the cutter. The authors found that all materials retained their strength with indentations of up to 400 N, likely due to no ring crack or damage of any form being evident at this load. At 1000 N, the fine-grained material experiences a drop in strength of around 65% to a strength below that of the coarse-grained material (Figure 15).

One possible explanation for this is the initial flaw size distribution. If the flaws do scale proportionally with the grain size as suggested by Petrovic et al. [38] then the lower initial strength of the coarse-grained material can be explained by larger initial flaws. After Hertzian indentation, the ring crack introduced could be much larger than the flaws in the fine-grained material but of a similar size to those already present in the coarse-grained PCD. This would then explain why the strength of the fine-grained material drops dramatically but that of the coarse-grained remains roughly constant – essentially the fine-grained material is much stronger to start with and therefore is more sensitive to damage on the scale of that caused by Hertzian indentation. However, without a measure of the initial flaw size distribution, it is impossible to tell if this explanation is correct.

Wear testing

The vast majority of the applications of PCD involve wear between two bodies; drilling of rock, machining of metal, wood and composites to name the most common. In light of this, an understanding of how PCD behaves when it is subjected to wear is highly important if we are to improve its wear resistance.

Drilling of rock

Hibbs and Lee conducted a study into the wear mechanisms that operate when PCD is used in a rock removal process [57]. They used very coarse-grained PCD, approximately 100-μm grain size, to cut a helical groove in a mortar rock cylinder containing hard particles of quartz. Figure 16 shows the two distinct wear morphologies that they found on the surface of the PCD after testing. The authors put forward an interesting hypothesis explaining the two differing morphologies. During sintering, the diamond grains undergo a large amount of plastic deformation due to the effect of high pressure combined with high temperature. It is suggested that this plastic deformation prevents easy cleavage of grains due to the disruption it causes to the lattice resulting in the chipping of small pieces of diamond from grains as seen in Figure 16(a). However, some grains may not be plastically deformed to the same degree by the chance arrangement of grains around them causing a bridging effect and preventing them from experiencing the same high stresses. These grains then retain their easy cleavage planes and can fracture transgranularly as seen in Figure 16(b). As there have been no studies on the effect of plastically deforming diamond on its fracture behaviour to the author’s knowledge, this remains an interesting hypothesis to test.

Another significant conclusion from their work is with regard to the wear resistance of the grain boundaries. The authors found that grains contained heavy scratch damage in the grinding direction. Cobalt pools had been preferentially removed from the surface layer but grain boundaries neither preferentially fragmented even when they were very close to the active cutting edge (Figure 17(a)) nor were preferentially abraded even when they were parallel to the cutting direction (Figure 17(b)). This gives further evidence that the grain boundaries do not represent a significantly weaker link as part of the PCD micro-structure even if the cobalt pools along them are removed, contrary to what many authors have suggested. When a large area was found to have broken away, a particularly large grain or evidence of poor intergranular bonding was often found showing that inhomogeneities in the sintering of PCD may act as critical flaws for fracture.

Lin et al. [58] identified three mechanisms of wear in a detailed study of cutters damaged in both laboratory rock-cutting tests and field operations. ‘Smooth wear’ occurred where diamond grains appear to have been polished to a flat surface with no significant fracture occurring. This type of wear was predominantly
seen in cutting through strong, abrasive, homogeneous rock where the drill head had a large number of bits distributed over its surface. Grooves oriented in the direction of cutting were also observed similar to those shown in other work [59, 60]. The rock being cut contained hard quartz particles but given the considerably higher hardness of diamond even to a very high temperature [61, 62], these grooves could not have been caused in a typical abrasive manner. The authors suggested that a combination of thermal and mechanical load on the diamond must have allowed either oxidation or graphitisation to occur causing these grooves to form.

The second wear morphology was labelled ‘micro-chipping’; although it may sound similar to one of the morphologies shown in Hibbs and Lee’s work, it can be seen by looking at Figure 18 that it is in fact quite different. This wear mechanism is characterised by small flakes of material chipping off where the plane of fracture is roughly parallel to the direction of cutting. The mechanism only appeared when cutting rock with a higher uniaxial compressive strength but low abrasiveness and after varying the cutting depth a number of times. The authors therefore concluded that this morphology was caused by some thermal or mechanical fatigue mechanism, citing work by Dunn and Lee who showed that the fracture stress of PCD cutters was significantly lowered by cyclic loading [63]. They noted that a higher negative rake angle (associated with higher forces in the direction of the cut) resulted in worse damage by microchipping. The mechanism proposed is that the cracks develop from tensile failure of the top face of the cutter. With a higher negative rake angle (i.e. the axis of symmetry of the cutter moving further away from parallel with the direction of cut) there is less material able to support the cutting edge and so larger tensile stresses are present on the cracks that develop here leading to increased microchipping.

The final morphology seen was that of gross fracturing. This involved large chunks of the cutter being removed instantaneously and represents the most problematic wear/fracture mechanism for drilling operations. Gross fractures were found to initiate on the curved surface of the PCD cutter rather than on the top face. Given that this type of fracture often occurred slightly after a change in rock formation during drilling [64], the authors suggested that it was as a result of mechanical overloading. To test this hypothesis, they loaded cutters against a steel platen in a similar geometry to that of rock cutting. They found that this loading condition caused fractures very similar to the ‘gross fractures’ seen in operationally used cutters and that by using a harder steel platen (causing an increased resistance of grain boundaries to fracture and abrasion after rock drilling shown by Hibbs and Lee [57]. Panel (a) shows that a grain boundary has not preferentially fractured despite being close to the cutting edge and panel (b) shows that there is no preferential abrasion of a grain boundary parallel to the cutting direction.

Figure 17. Resistance of grain boundaries to fracture and abrasion after rock drilling shown by Hibbs and Lee [57]. Panel (a) shows that a grain boundary has not preferentially fractured despite being close to the cutting edge and panel (b) shows that there is no preferential abrasion of a grain boundary parallel to the cutting direction.

Figure 16. Wear morphologies seen after rock drilling by Hibbs and Lee [57]. Panel (a) shows sub-grain chipping and panel (b) shows transgranular fracture of multiple grains.
stress at the cutter’s edge), gross fracturing occurred at lower loads.

Yahiaoui et al. [65] performed a study comparing the wear resistance of fine (6 ± 1 \( \mu \)m) and coarse (17 ± 4 \( \mu \)m)-grained PCD sintered using two HPHT regimes. A standard regime (HPHT1) and one with a higher pressure but shorter dwell time. In the introduction, the author states that it is ‘well known by the manufacturers [that] PDC cutters with a fine grain distribution are more resistant to abrasion than cutters with coarse grains’ citing a review paper on the current state of PDC bit technology [15]. However, the results of the study did not agree as it was found that the coarse-grained PCD was more wear resistant than fine grained. The authors did not look for a mechanistic difference in the wear behaviour and suggested that other microstructural parameters might be more important to the wear behaviour than simply average grain size.

The influence of binder removal on the wear behaviour of PCD was examined by Liu et al. [27]. The cobalt was removed from PCD to various depths by electrolysis and the wear ratio (ratio between amount of work piece removed and amount of cutter worn away) evaluated for each removal depth over a fixed sliding distance. It was found that the wear ratio of the cutter increased monotonically up to a removal depth of around 200 \( \mu \)m showing that the removal of the binder phase increases the wear resistance of the cutter. This was also seen in a visual comparison with binder-removed cutters having a far smaller wear scar than those with residual metal still present. Deep grooves were noted on the wear scars of the binder-removed cutters. Although the authors did not suggest a mechanism by which this could have happened, they suggest that it may constitute a self-sharpening behaviour as these grooves reduce the contact area between the cutter and the work piece, reducing heat and friction at the contact and hence improving the cutting efficiency. XRD measurements showed that there was a complete absence of graphite on the wear scars of the binder-removed material. This potentially corroborates Lin et al.’s suggestion that graphitisation is a significant source of wear and shows that catalysis by the binder phase is the pathway by which this happens. Another possible explanation is that the removal of the binder reduces the friction of the contact [66] and hence inhibits graphitisation by a virtue of a lower temperature interface.

In a study of microstructural characterisation of the thermal degradation of PCD after milling of granite [67], Westraadt et al. found that volume expansion due to graphitisation was a major contributor to failure of the cutter. This finding contradicts much of the industrial wisdom that it is the thermal expansion of the cobalt that causes thermally induced cracking. They found through comparison with static annealing experiments that the temperatures generated at the cutting interface must have exceeded 800°C which is sufficient to cause the cobalt-catalysed conversion of diamond to graphite. Another interesting finding of their work is that much of the cracking they observed was intergranular. This suggests that there may be a different fracture mechanism activated when PCD is used for drilling applications compared to pseudo-static mechanical tests at room temperature. If cobalt along the grain boundaries is causing conversion of diamond to graphite at high temperature, resulting in it expanding, it seems logical that the material would fracture along these grain boundaries rather than transgranularly as is seen at low temperature.

**Machining of titanium**

A growth area for PCD is in the machining of high-strength titanium alloys that are particularly difficult to machine with traditional carbide and high-strength steel tooling. Turning tests of five different diamond grain sizes on a billet of titanium found that the tool life increased as grain size increased from 1.3 to 14 \( \mu \)m [68]. Although, the largest grain size of 39 \( \mu \)m performed the worst by far, experiencing grain-pullout and ‘sub-grain transgranular fracture’.

A study was conducted to measure the effect of cutting speed on the wear behaviour of PCD when used to turn Ti–6Al–4V titanium alloy [69]. It was found that the predominant wear mechanism depended on the cutting speed used. At the lower-cutting speed, the wear mechanism was described as attritional with a rough, discontinuous wear surface and a complete lack of adhered material to the tool. At the higher-
cutting speed, a smooth wear surface was found characteristic of diffusive/dissolutional wear. The presence of adhered material on the cutter would seem to support this. Finite element modelling simulations were used to calculate the expected temperatures at the cutting interface, the results of which predicted a temperature of 897°C for the lower-cutting speed and 991°C for the higher. The authors point out that a temperature difference of only 100°C is not enough to explain a difference in diffusivity so large that it would cause the completely different wear mechanisms seen between the two cutting speeds. It was therefore suggested that a temperature-induced phase change in the chip material caused the diffusivity of tool elements into the titanium to increase by orders of magnitude. This hypothesis appears to be backed up by the alloy’s phase diagram (taking pressure into account). A secondary argument is that the HCP α phase of titanium contains only six facile slip systems, whereas the BCC β phase has 12. This means that at the lower-cutting speed where the chip material is predicted to be of the α phase, the chip will not plastically deform as easily, increasing the stress on the cutting edge and causing it to fracture more frequently.

**Machining of wood composites**

Bai et al. [70] performed tests involving the machining of laminate wood composites using four different PCD grain sizes (2, 10, 25 and 75 μm). It should be noted that these are the average sizes of the original diamond grains not a measurement of the sintered grain size. Micro-cracks were observed in the 10-μm grain size material (Figure 19). It was suggested that it is the formation of these microcracks that are responsible for the ‘rupture and disintegration of the PCD tool’. The authors also suggest that these microcracks can propagate along grain boundaries. It may be that this occurs in larger-grained microstructures, being responsible for the often mentioned ‘grain pullout’ but it has been shown on numerous occurrences that the grain boundaries do not act as a preferential path for failure so this statement seems unlikely without any evidence provided to support it. An interesting claim is made that the ‘inherent cleavage cracks generated in HTHP synthesis also contribute to the formation of cleavage wear’, however, as the paper cited against this claim is in Chinese [71], further investigation has proved difficult. The final conclusion of this paper is that middle-sized diamond grains (10 μm) have the highest wear resistance over both large and small grain sizes, further highlighting the error in what is ‘known by all manufacturers’.

Philbin and Gordon [11] analysed the wear mechanisms that operated when cutting two different types of wood composite with PCD. They found that when cutting a more homogeneous material, abrasive wear was dominant, with voids where individual grains had been pulled out being seen. When a material with a protective, hard layer was cut, multi-grain chips were observed at the cutting edge instead.

A summary of all of the wear mechanisms described herein is given in schematic form in Figure 20.

**Summary**

It is clear from the literature presented in this document that there are still many fundamental aspects of the relationship between microstructure and mechanical behaviour of PCD that are poorly understood. Some of the assumptions made to help understand this behaviour are based either on questionable conclusions or long held industrial wisdom, both of which are unhelpful in furthering the scientific understanding of the material.

A better explanation for the strength-grain size behaviour of PCD is suggested based on consideration of the flaw size distribution, but there is currently little existing literature to corroborate this. There appears to be consensus between authors that increasing the grain size of PCD increases its toughness, counter to the effect seen in many traditional ceramics, however, no obvious reason has been discovered for this phenomenon.

Many authors present micrographs of transgranular fracture occurring in PCD yet when explaining the fracture behaviour, persist in describing grain boundaries as weak paths for fracture. It is clear that these two arguments are incongruent and that a better understanding of the fracture behaviour of PCD is required. In a number of the wear studies, different
wear morphologies such as smooth wear, grain pullout and microchipping are seen under different conditions. Intergranular fracture is almost never seen in pseudo-static testing but frequently appears in the results of wear tests. It is suggested that the main driving factor behind this phenomenon is the effect of temperature on the material at the grain boundaries which is as yet relatively unstudied.

The binder phase clearly has an important effect on the mechanical properties of PCD. Marro et al. showed that it had a local toughening effect as cracks passed through it. However, the collated data in section 'Fracture toughness' suggests that increasing the overall binder content of a PCD material reduces its toughness. Leaching appears to greatly improve the wear behaviour of PCD while also moderately degrading its toughness which appears to be contradictory to the previous statement. The interaction of cracks and the binder phase is clearly complicated, but it is difficult to vary the binder content of PCD independently of the grain size so a more direct way of analysing this interaction is required.

Finally, it has been suggested that the plastic deformation of diamond grains that occurs during sintering could have an effect on the mechanical properties of PCD. Some grains may remain undeformed if bridging

Figure 20. Schematic representation of all of the wear mechanisms of PCD described above. (a) Sub-grain microchipping due to plastic deformation blocking easy cleavage planes [57], (b) bridging of individual grains prevents plastic deformation allowing single grains to fracture along easy cleavage planes [57], (c) smooth wear caused by dissolution of carbon into the work piece and/or chemical attack [58], (d) microchipping caused by tensile cracks opening up on the PCD table [58] (figure taken from reference) (e) gross fracturing caused by excessive normal force on the cutter [58] (figure taken from reference), (f) pullout of an entire grain caused by a weakening of the grain boundary [68].
by neighbouring grains occurs allowing these deformation-free grains to act as critical flaws. However, little research has been conducted investigating this effect.

By investigating these gaps in the literature, a much more in-depth and accurate understanding of how the microstructure of PCD affects its mechanical properties can be gained. This will allow continued development in the field of high performance cutting tool materials creating technological advancements in the aerospace, automotive, resource exploration and transportation industries of the same or greater magnitude as when PCD first entered the market.

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