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**Ball Indentation Fatigue of
Powder Metallurgy Steels**

A J GANT

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Dr Andrew J Gant
Engineering and Process Control Division

ABSTRACT

Commercial powder metallurgy (PM) steels produced from water atomised highly alloyed powders which are pressed and gas sintered in mesh belt furnaces find ready application for use as automotive valve train components; they can be easily mass produced and require relatively little machining after being sintered/heat treated. Material specifications for PM steels for automotive applications are becoming increasingly exacting as operating conditions become more arduous and also the design lifetimes of components become longer. In the context of the latter there is at present a dearth of data on PM steels subjected to dynamic loading, and especially under relevant loading modes; hence the present study.

This NPL open report first sets the context in which these materials are used, outlines their material property requirements and briefly describes typical valve and valve seat insert compositions. An introduction to contact between a sphere and plane is also given, with both elastic and plastic deformation of the latter. Prior to conducting dynamic loading experiments, static and quasi-static material properties were measured in order to determine the dynamic loading regimes to be used. Response to dynamic loading was assessed by measuring indentation depth and examining the polished surfaces of the samples. Further examination was conducted on one selected material by sectioning vertically through indentations, polishing and then inspecting damage under the indentations by scanning electron microscopy.

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National Physical Laboratory
Hampton Road, Teddington, Middlesex, TW11 0LW

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Approved on behalf of the Managing Director, NPL,
by Dr M G Gee, Knowledge Leader, Materials Performance Team
authorised by Director, Engineering and Process Control Division

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1 VALVE AND VALVE OPERATION CONDITIONS

The requirements placed upon valves and valve seats as a result of their operating environment has been detailed by several authors [1-4]. The valve train environment has become more severe in recent years due to several factors, necessitating the development of new materials which will withstand the increasingly hostile operating conditions. Most reasons for these changes fall into two categories – environmental concerns and the demand for greater economies. In practical terms this means higher output per unit cylinder capacity [2], the use of turbocharging and of alternative fuels such as LPG or methanol [5]. Lower specific fuel consumption (e.g. in gasoline engines from near stoichiometric fuel/air ratios), faster operating speeds, lower oil consumption, reduced maintenance and greater durability to extend engine life are also being demanded [6].

Exhaust operating temperatures are usually within the range 425-750 °C [1]. There is some lubrication due to the presence of combustion products [7], but it is poor [1,7].

The valve/valve seat environment is also corrosive due to the presence of hot exhaust gases. Sulphur and ash are also major contributors in low grade diesel fuel [8]. Vanadium and sodium present in fuel also degrade coatings by forming low melting point salts (complexes of sodium and vanadium oxides) [9,10]. $\text{Al}_2\text{O}_3\text{-SiO}_2$ from catalytic cracking can also cause erosion. The nomenclature for the valve and valve seat are shown in Figure 1. Valve and valve seat degradation mechanisms are summarised in Figure 2.

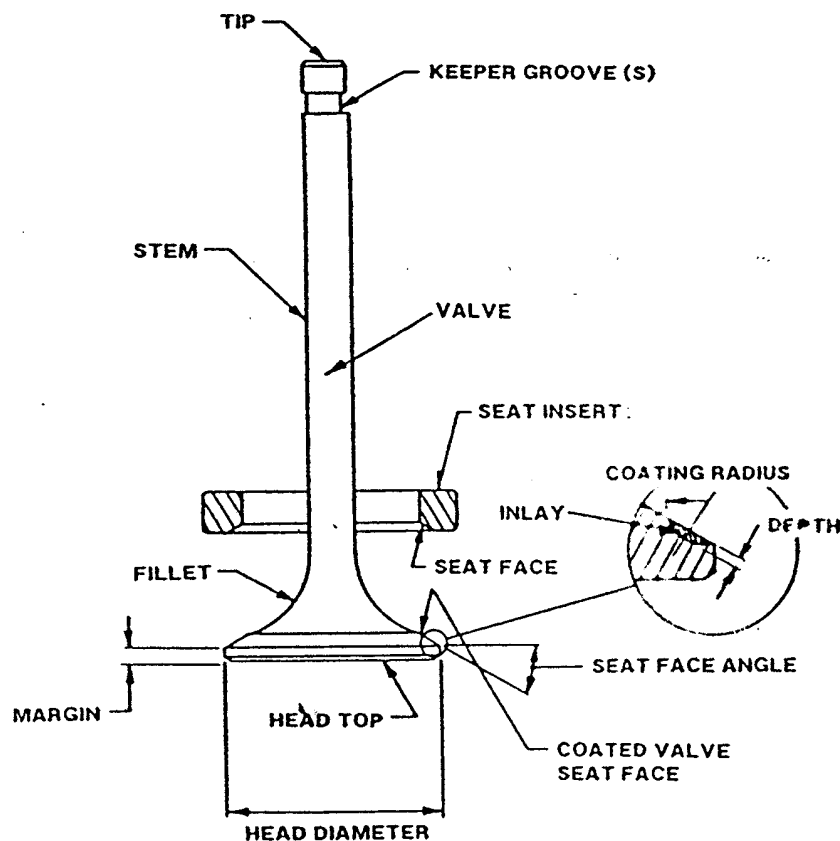


Figure 1: Nomenclature for the valve and valve seat.

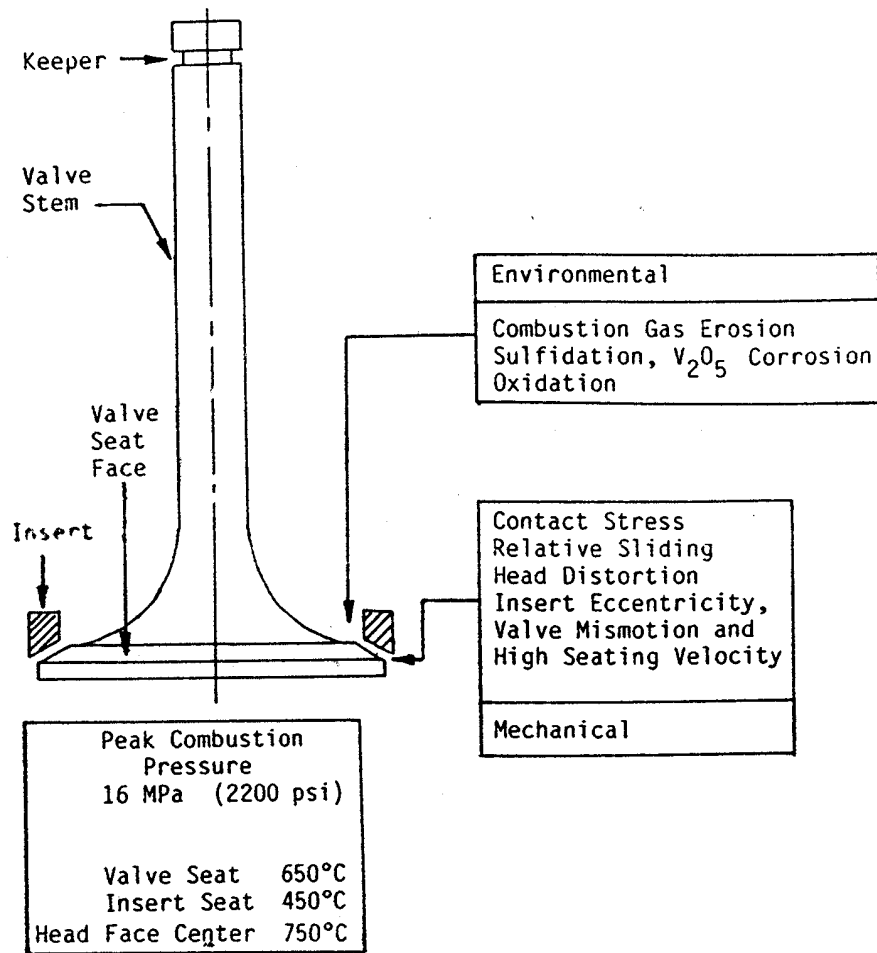


Figure 2: Valve degradation mechanisms.

2 VALVE/VALVE SEAT INSERT PROPERTY REQUIREMENTS

Valve material properties needed to resist the deformation mechanisms in Figure 2 are outlined below in sections 2.1-2.3, with particular emphasis being placed upon the head/seat area. Forces acting on the valve seat area are shown in Figure 3.

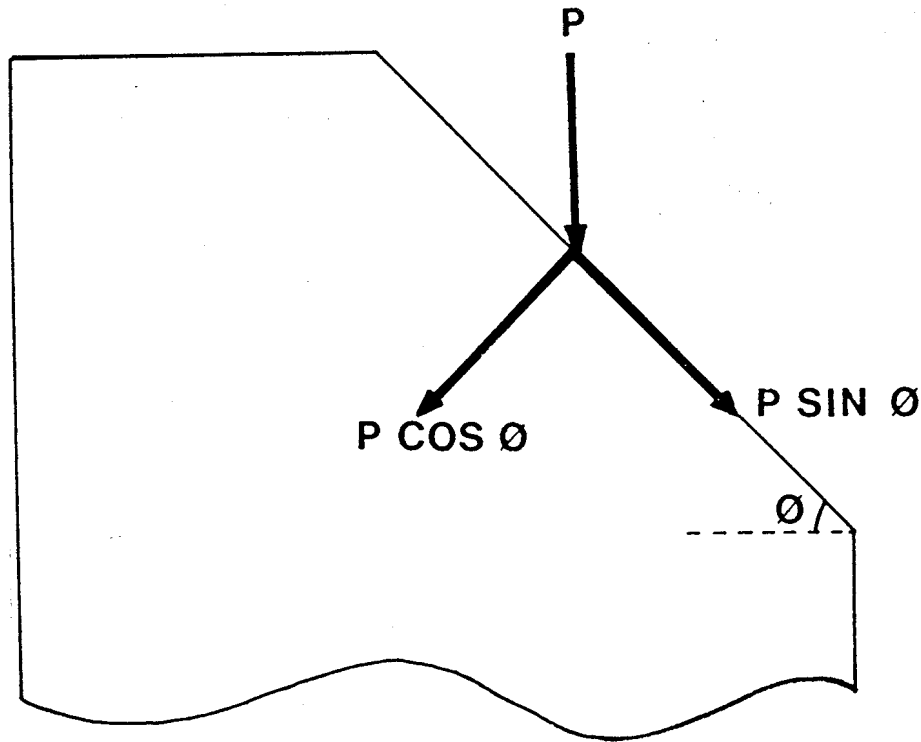


Figure 3: Forces acting on the valve seat insert face.

2.1 CREEP (DISTORTION) RESISTANCE

The valve head must possess good creep resistance to prevent it from permanently deforming (“dishing”) when operating at high temperatures and under alternating loads; this is especially true for the underhead radius and the seat area [11]. In gasoline engines, the maximum temperatures occur at the exhaust valve stem blend and underhead radius where exhaust gases impinge, resulting in temperatures between 650 and 800°C. The valve material should have a moderate coefficient of thermal expansion to limit thermal stresses due to the large temperature gradient across the head, but a high thermal conductivity to aid rapid dissipation of heat from the head to the seat. Stresses vary with location in the valve with the highest stresses occurring in the valve underhead radius and in the seat area. In gasoline engines stresses in the exhaust side can be in the range 35-70 MPa. Seat insert distortion should be minimised as it can raise stresses in that region ten-fold due to localised leakage, initiating valve burning, and have a pronounced effect on valve head temperatures [12,13]. In faced valves seat distortion can cause aggravated thermal stresses, causing the face to crack and severe guttering to occur behind the hard face surface. A major cause of seat distortion is excessive temperatures due to insufficient cooling as a result of poorly designed coolant channels or bad foundry practice.

2.2 WEAR RESISTANCE

Hot strength and hardness (these often being used as a measure of valve seat face peening and wear resistance) should be adequate to resist cupping of the head and rapid seat wear. Indentation (“brinelling”) of the valve seat can result from excessive temperatures, poor hot hardness, or entrapment of hard deposits [1]. Material removal

(wear) at the valve/insert interface is generally termed “recession” and is classified by van Dissel [6] as one of two principal types of valve distress; the other being corrosion. Seat wear occurs due to the relative motion between valve and seat on opening and closing. This relative motion takes the form of impact and sliding (rotation) [14], though others [15] suspect frictional sliding to be the main cause. Both adhesive and abrasive wear modes and contact fatigue can occur [1,6].

2.3 FATIGUE RESISTANCE

The valve should have good fatigue resistance (especially in the valve head and stem blend regions) to combat repeated cyclic stress loading; “breakage”, or catastrophic failure of the valve is most commonly due to high cycle fatigue [13]. Fatigue failures are progressive, beginning with crack initiation followed by crack growth under the action of a sufficiently large number of cycles of the alternating stress. Final fracture occurs after the crack has become large enough such that the remaining cross section can no longer carry the load. Exhaust valves are particularly susceptible to this type of failure. Seat distortion creates bending stresses in the underhead, stem-blend area of the valve due to misalignment of the valve stem relative to the seat. These bending stresses coupled with the cyclic nature of the valve motion ultimately lead to a fatigue failure.

In gasoline engines stem or stem-blend failures usually occur, as this is where the maximum valve temperature occurs (being the point of exhaust gas impingement) and hence the weakest point, but diesel engine exhaust valves are designed with a larger fillet radius to cope with the higher combustion pressure and increased gas flow. As a result, the head thickness tends to be the critical factor and chordal failure (in which a chordal segment breaks off due to impact fatigue) fractures are observed [13].

3 VALVE MATERIALS

Metallic exhaust valve materials can be classified into three basic groups; silicon-chromium steels, austenitic chromium-nickel steels and nimonic nickel base alloys. Early valves were of silicon-chromium steel, typically of the following composition: 0.8% C, 0.4% Mo, 1.3% Ni, 2% Si and 20% Cr. Such materials have a maximum operating temperature of 650°C. With improved engine performance, the need to operate at higher speeds and temperatures was met by austenitic steel valves. By increasing the nickel content to 12%, the corrosion resistance of such steels was adequate for the conditions they experienced. Typical of this group of steels was “21-12”; having a composition of 0.25% C, 1.5% Mn, 1% S, 12% Ni and 21% Cr. A further development of this type of valve was achieved by incorporating 0.4% nitrogen into the steel, increasing manganese to 9% at the expense of the nickel content: “21-4N”. With even higher temperatures being required for valves to operate as a result of leaner air-fuel mixtures under partial load, nimonic (nickel base) alloys such as “80A” were employed.

The hot strength and hardness, fatigue and corrosion resistance of nimonics are superior to austenitic steels, but they are more expensive. Since the valve head is exposed to the most arduous conditions and the stem less so, a solution meeting the material requirements at the head end but yet being relatively cheap was the two-piece valve having a nickel base alloy head and a steel stem joined by friction welding. Increasing demands on erosion, friction and wear resistance in exhaust valves lead to the

increasing use of Stellite facing [16], a common grade being Stellite #3 – a cobalt based alloy with additions of chromium, tungsten and nickel. Recently however, Savel [17] has reported promising results of unfaced one-piece valves being used in both the inlet and exhaust positions.

4 VALVE SEAT INSERTS

Valve seat inserts (VSIs) were first developed for spark ignition car engines. Formerly, grey cast iron for cylinder head castings possessed adequate properties for use as valve seats [2]. The free graphite in grey cast iron imparted wear resistance and cast iron had a relatively high thermal conductivity. The trend towards higher power outputs per unit cylinder capacity, turbocharging and alternative fuels have made conditions in the valve/valve seat area more arduous. There has also been a concomitant increase in the use of aluminium heads in spark ignition car engines to reduce weight and increase heat transfer. This latter development prompted the use of the first VSIs produced by a powder metallurgical route [2,18]. The latter offers several advantages over the cast insert; costs are reduced due to lower material wastage, a higher production rate can be achieved and less energy is used in production. A more refined microstructure can also be achieved by the PM route (aiding machinability and hence reducing tooling costs) and hard phases and/or solid state lubricants can be incorporated.

Conditions under which the valve and valve seats operate in diesel engines are much more severe than gasoline engines. On the inlet side typical seat temperatures are approximately 300°C (50°C higher than for an SI engine) [19,20], but more importantly, the diesel inlet seats experience “dry” conditions as only air is inducted [21].

In the past, PM alloys were only used for light duty valve seats, with cast superalloys being used for heavy duty applications [22]. The latter are however expensive and difficult to cast; the PM route can frequently achieve a material yield of over 90% - casting processes rarely reach this level and can fall as low as 40% [2].

Light duty exhaust valve seat applications require only low alloy PM materials; these typically being iron-based material containing a minimum of approximately 1.5% alloying additions (molybdenum, manganese and nickel). Such materials are used in the sintered condition. Microstructures consist mainly of tempered martensite with some pearlite and free, well-dispersed ferrite. Typical matrix hardnesses range from 37 HRc to 43 HRc.

Medium duty valve seats require more highly alloyed materials; typically iron-based with additions of nickel, molybdenum, cobalt, manganese and chromium. Such materials have a martensitic matrix with a fully continuous carbide particle network.

Heavy duty applications call for excellent wear resistance and mechanical properties at elevated temperatures which may necessitate a material not only to possess a martensitic matrix, but also to feature a fine dispersion of carbides, entailing the use of yet more highly alloyed materials [22]. These are again typically iron-based materials, but with additions of molybdenum, tungsten, chromium, vanadium and nickel. Matrix hardness ranges from 48 HRc to 56 HRc.

5 CONTACT BETWEEN SURFACES

5.1 ELASTIC DEFORMATION BETWEEN A SPHERE AND A PLANE

In the field of contact mechanics, the elastic contact between a sphere and a plane has been particularly closely studied [23]. When a sphere of an elastic material is pressed against a plane (Figure 4) under a normal load w , contact will occur between the two over a circular area of radius a , given by the following equation due to Hertz [24], first published in 1882:

$$a = \left(\frac{3wr}{4E} \right)^{\frac{1}{3}} \quad (1)$$

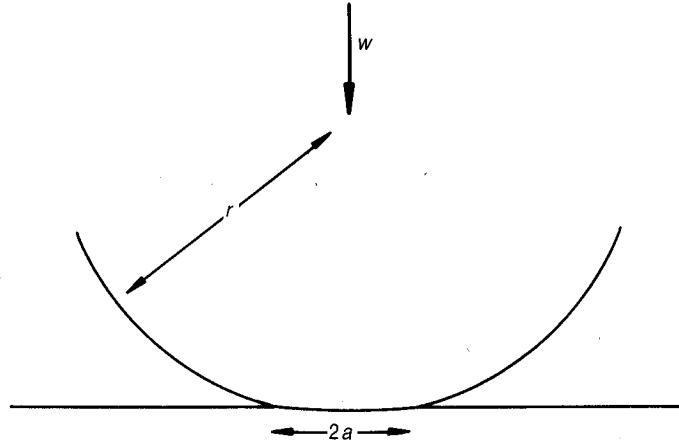


Figure 4: Elastic deformation of a sphere of radius r , pressed against a plane surface under a load w .

Here r is the radius of the sphere and E^* is an elastic modulus which depends on Young's moduli, E_1 and E_2 and on the Poisson's ratios ν_1 and ν_2 , for the materials of the sphere and of the plane in the following way:

$$\frac{1}{E^*} = \frac{(1-\nu_1^2)}{E_1} + \frac{(1-\nu_2^2)}{E_2} \quad (2)$$

The area of contact between the sphere and the plane, πa^2 , is then given by:

$$\pi a^2 \approx 0.83\pi \left(\frac{wr}{E^*} \right)^{\frac{2}{3}} \quad (3)$$

For this case, in which deformation is purely elastic, the area of contact is therefore proportional to $w^{\frac{2}{3}}$. The mean pressure (normal stress) P_{mean} over the contact area is $w/\pi a^2$, and thus varies as $w^{\frac{1}{3}}$. The stress is not uniform over the circular area of

contact, but has a maximum at the centre and falls to zero at the edge. The distribution of normal stress is shown in Figure 5; the maximum stress, at the centre of the contact circle is $3/2$ times the mean pressure.

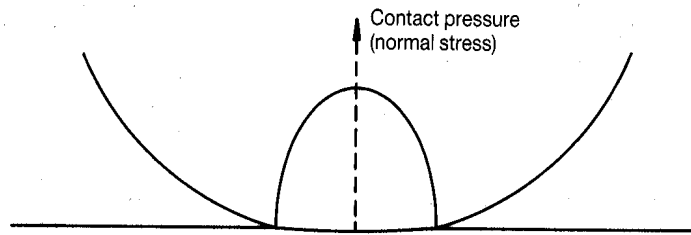


Figure 5: The distribution of normal stress (contact pressure) under a sphere loaded elastically against a plane.

5.2 PLASTIC DEFORMATION BETWEEN A RIGID SPHERE AND A PLANE

As the normal load between the sphere and the plane is increased, one or other of the components may start to deform plastically. The situation may be simplified by considering the case in which the sphere is assumed to be rigid, and plastic flow confined to the plane.

The indentation of a plastic half-space by a rigid sphere has been extensively investigated in connection with indentation hardness testing. Hertz's analysis of the elastic stress field due to a spherical indenter on a flat surface shows that the maximum shear stress beneath the indenter occurs at a depth of about $0.47a$ (where a is the radius of the contact circle). Plastic flow first occurs at this point when the yield criterion is satisfied. If the Tresca criterion is applicable, this will be at a maximum shear stress of $Y/2$ where Y is the uniaxial yield stress of the material. For a material in which Poisson's ratio has a value of about 0.3, the shear stress at a depth $0.47a$ beneath the sphere is 0.47 times the mean contact pressure. Plastic deformation therefore initiates at a mean contact pressure of $1.1Y$.

As the normal load is further increased, the zone of plastic deformation extends from beneath the indenter until it eventually reaches the surface. At this point, which may be shown theoretically to occur in a metal at a load some 50 to 100 times the load at which plastic flow initiates, the contact area is still extremely small, with a radius typically less than 1% of the radius of the sphere. The mean pressure over the contact area has risen by this stage to about $3Y$ and remains at nearly the same value for subsequent increases in load. The independence of the mean contact pressure from the load once full plasticity has been reached, and the constant proportionality between that pressure and the yield stress of the indented material provide the basis for indentation hardness testing. The constant of proportionality between indentation pressure and yield stress depends on the geometry of the indenter, but is never far from the value of 3 found for the case of a sphere. Some care should be exercised when obtaining estimates of yield strength from hardness values as the following will show:

The 0.2% offset yield strength can be determined [25] with good precision from Vickers hardness measurements according to the relation:

$$\sigma_0 = \frac{DPH}{3} (0.1)^{n'-2} \quad (4)$$

where σ_0 = the 0.2% offset yield strength, kgf mm⁻² (=9.81 MPa)
 DPH = the Vickers hardness number
 $n' = n + 2$ = the exponent in Meyer's law

Meyer's law is an empirical relation between the load and size of the indentation [26]:

$$P = kd^{n'} \quad (5)$$

where P = applied load, kg
 d = diameter of indentation, mm
 n' = a material constant related to strain hardening of the metal
 k = a material constant expressing resistance of the metal to penetration

Fully annealed metals have a value of n' of about 2.5, while n' is approximately 2 for fully strain-hardened metals. It can then be seen that for a Meyer's law exponent of 2 that 0.2% offset yield strength is simply one third of the Vickers hardness number.

6 MATERIALS INVENTORY

Five powder metallurgy (PM) steels were supplied in the form of annular specimens by Federal Mogul Sintered Products Ltd, Coventry. Details of the materials are outlined in table 1.

Table 1: Details of PM steel samples supplied by Federal Mogul Sintered Products.

Material Code	Details
3000	100% high speed steel, pressed to a green density of ~6.45 g/cc. Infiltrated with Cu to achieve a final density of ~8.0/8.1 g/cc; Cu content of the product ~25%
32xx	The blend is ~50% high speed steel + ~50% unalloyed Fe; pressed to ~6.6/6.7 g/cc. Infiltrated with Cu to achieve a final density of ~7.8/7.9 g/cc. Cu content of the product ~15/17%
1200	Fe + ~4/5% Cu + C blend; pressed to ~6.6/6.7 g/cc. Infiltrated with Cu to achieve a final density of ~7.8/7.9 g/cc. Cu content of the product ~15%.I
1601	Fe + ~1% C + ~14/15% Cu, obtained by a non-infiltration process; Density ~7.0; oil absorbed during density determination ~0.5%.
M3CU	High speed steel + ~15% Cu, obtained by a non-infiltration route. Density ~7.1; oil absorbed during density determination ~0.5%, but if the oil impregnation is done under reduced pressure, % oil absorbed ~1.5%. For routine laboratory tests, oil impregnation for the density determination is done at atmospheric pressure

7 MEASUREMENT OF MATERIAL PROPERTIES

7.1 STATIC AND QUASI-STATIC PROPERTIES

In order to determine what indentation fatigue regimes were to be used, a portfolio of static mechanical properties was collated prior to beginning the fatigue tests. Static tests were carried out both at Federal Mogul Sintered Products Ltd, Coventry, and also in-house at NPL.

0.2% Offset Compressive Yield Strength and Young's Modulus

Data were supplied by Federal Mogul and are shown in table 2:

Table 2: Compressive yield strength and Young's modulus values supplied by Federal Mogul Sintered Products.

Material Code	Heat Treated Condition	0.2% Offset Compressive Yield Strength, GPa	Young's Modulus, GPa
3000	Cryogenic quench - 120°C + 585°C temper	1.47	190
32xx	Cryogenic quench - 120°C + 600°C temper	1.10	185/200
1200	As sintered	0.64	170
1601	As sintered	0.68	140
M3CU	Cryogenic quench - 120°C + 585°C temper	1.21	132

At NPL, Vickers hardness was measured with a 30kg load and Rockwell hardness was measured on the "C" scale; i.e. HRc. The results are in table 3 (mean values of five indentations per sample for both test formats are given).

Table 3: Mean HV30 and HRc hardness values

Material Code	Heat Treated Condition	Mean Vickers Hardness, HV30	Mean Rockwell Hardness, HRc
3000	Cryogenic quench - 120°C + 585°C temper	427.4	43.8
32xx	Cryogenic quench - 120°C + 600°C temper	458.3	44.1
1200	As sintered	272.5	27.2
1601	As sintered	267.9	24.4
M3CU	Cryogenic quench - 120°C + 585°C temper	322.6	34.2

It can be seen that if a Meyer's law exponent is assumed then the 0.2% offset yield strength can also be estimated from the Vickers hardness values. It is assumed that under the Vickers indenter that the metals in question are fully strain hardened, hence a Meyer's law exponent of 2. Estimated values (from HV30 measurements) and actual 0.2% offset compressive yield strength values are given in table 4.

Table 4: Comparison of measured 0.2% offset compressive yield strength values with values calculated from Vickers hardness measurements.

Material Code	Heat Treated Condition	Measured 0.2% Offset Compressive Yield Strength, GPa	Estimated 0.2% Offset Compressive Yield Strength From HV30 Values, GPa
3000	Cryogenic quench - 120°C + 585°C temper	1.47	1.40
32xx	Cryogenic quench - 120°C + 600°C temper	1.10	1.50
1200	As sintered	0.64	0.89
1601	As sintered	0.68	0.88
M3CU	Cryogenic quench - 120°C + 585°C temper	1.21	1.05

7.2 QUASI-STATIC BALL INDENTATION TESTS

Quasi-static ball indentation tests were performed on an Instron 4505 in compression mode at a crosshead speed of 0.2 mm.min⁻¹. The PM valve train materials were received in the form of ring shaped specimens. They were mounted in bakelite and metallographically polished with diamond aerosols to 1µm and given a relief polish using a 0.05µm non-crystallising silica suspension. Compression tests were carried out using WC-6%Co hardmetal balls of 8mm and 25mm diameters being pressed into the polished flat faces of the annular PM specimens. The WC-Co balls and PM specimens were compressed between hardened steel platens. Load-displacement traces were produced by a chart recorder.

7.3 ELASTIC CONTACT CALCULATIONS FOR WC-6%CO BALLS AND PM STEEL SPECIMENS

Radius of Contact Circle

As outlined previously, this is given by $a = \left(\frac{3wr}{4E^*} \right)^{\frac{1}{3}}$ (1)

where r is the radius of the sphere and E^* is an elastic modulus which depends on Young's moduli, E_1 and E_2 and on the Poisson's ratios ν_1 and ν_2 , for the materials of the sphere and of the plane in the following way:

$$\frac{1}{E^*} = \frac{(1-\nu_1^2)}{E_1} + \frac{(1-\nu_2^2)}{E_2} \quad (2)$$

Where the ball is denoted by ₁ and the flat is denoted by ₂.

Relationships between binder content, Young's modulus and Poisson's ratio have been derived and verified by experiment and are available in the literature [27]. For a 6% cobalt WC hardmetal, values of Young's modulus and Poisson's ratio are calculated as:

$$\begin{aligned}\nu_1 &= 0.228 \\ E_1 &= 617.5 \text{ GPa}\end{aligned}$$

The "composite" modulus of the ball/flat couple, E^* , can now be calculated and values for all five PM steels and WC-6%Co balls are shown in table 5:

Table 5: Ball/flat "composite" Young's modulus values; E^* .

Flat Material	Ball/Flat "Composite" Young's Modulus E^* , GPa
3000	158.1
32xx	154.9
1200	145.2
1601	124.5
M3CU	118.6

In order to take the analysis further, one must assume that only one of the components will start to deform elastically and then yield as the load is increased. Given the much higher elastic modulus (hence the ball is assumed to be rigid) and hardness of the ball, it is assumed that both elastic deformation and plastic flow are confined to the plane. If a Poisson's ratio of 0.3 is assumed for the plane, the shear stress at a depth of $0.47a$ beneath the sphere is 0.47 times the mean contact pressure. Plastic flow first occurs at this point when the yield criterion is satisfied; for the Tresca criterion, this will be at a maximum shear stress of $Y/2$, where Y is the uniaxial yield stress of the material. Plastic deformation therefore initiates at a mean contact pressure of $1.1Y$.

Now
$$P_{mean} = \frac{w}{\pi a^2} \quad (6)$$

Where w = applied load and a = radius of contact circle

Therefore
$$1.1Y = \frac{w}{\pi a^2} \quad (7)$$

$$w = \pi a^2 (1.1Y) \quad (8)$$

But if the onset of yielding is only just reached, the elastic contact equation will still apply:

$$a = \left(\frac{3wr}{4E^*} \right)^{\frac{1}{3}} \quad (1)$$

Thus
$$w = \pi \left(\frac{3wr}{4E^*} \right)^{\frac{1}{3}} (1.1Y) \quad (9)$$

$$w \left(\frac{3wr}{4E^*} \right)^{\frac{2}{3}} = \pi(1.1Y) \quad (10)$$

$$w^{\frac{1}{3}} = \left(\frac{3r}{4E^*} \right)^{\frac{2}{3}} \pi(1.1Y) \quad (11)$$

$$w = \left(\left(\frac{3r}{4E^*} \right)^{\frac{2}{3}} \pi(1.1Y) \right)^3 \quad (12)$$

7.4 CALCULATION OF ONSET OF PLASTIC DEFORMATION IN FLAT PM RINGS

The onset of plastic deformation in the Federal Mogul materials has been calculated for the case of both an 8mm and a 25mm diameter WC-6% Co ball being pressed into the flat using equation (12). The values calculated are shown in table 6.

Table 6: Calculated values for the onset of plastic deformation in the PM steels

Material Code	Calculated Compressive Force to Initiate Plastic Deformation Using an 8 mm Diameter WC-6% Co Ball, N	Calculated Compressive Force to Initiate Plastic Deformation Using an 25 mm Diameter WC-6% Co Ball, N
3000	189	1846
32xx	82	804
1200	18	180
1601	30	295
M3CU	187	1826

7.5 EXPERIMENTAL VALUES FOR ONSET OF PLASTIC DEFORMATION IN FLAT PM RINGS

Table 7: Predicted and actual compressive forces to initiate plastic deformation in flat PM rings using an 8mm diameter WC-6% Co ball.

a) 8 mm Diameter Ball

Material Code	Predicted Compressive Force to Initiate Plastic Deformation, N	Actual Compressive Force to Initiate Plastic Deformation, N
3000	189	112.5
32xx	82	42.5
1200	18	41.3
1601	30	47.5
M3CU	187	45

Table 8: Predicted and actual compressive forces to initiate plastic deformation in flat PM rings using an 25mm diameter WC-6% Co ball.

b) 25 mm Diameter Ball

Material Code	Predicted Compressive Force to Initiate Plastic Deformation, N	Actual Compressive Force to Initiate Plastic Deformation, N
3000	1846	190
32xx	804	110
1200	180	40
1601	295	25
M3CU	1826	30

7.6 INDENTATION FATIGUE REGIME

Fatigue tests were conducted using an Instron 8872 servohydraulic fatigue machine under the regime shown in table 9:

Table 9: Details of indentation fatigue regime.

Maximum Load	3.00kN (compressive)
Minimum Load	1.5kN (compressive)
Mean Load	2.25kN (compressive)
Amplitude	0.75kN (i.e. R=0.5)
Peak to Peak Frequency	10Hz
Test Durations (Number of Cycles)	10, 100, 1000, 10 ⁴ , 10 ⁵ , 10 ⁶

The fatigue tests were conducted using 25mm diameter WC-6% Co balls identical to those used in the second series of static tests. A fresh ball was used for each individual fatigue run. The tests were stopped automatically once the preset test duration was reached. Both load and displacement trip settings were programmed into the Instron 8872 control panel so that a test would be automatically stopped should there be a load or displacement excursion (due for instance to failure of the ball or sample). All tests ran to their preset durations; none were stopped prematurely due to load or displacement excursions.

7.7 ASSESSMENT OF MATERIAL DEFORMATION AND DEGRADATION DUE TO IMPACT FATIGUE

Material deformation due to indentation fatigue was measured using a Nikon MM-60 Measuring Microscope (optical microscope with a travelling x-y table and a 3D displacement measuring system). 3D profilometry was attempted using an Olympus LEXT OLS3000 confocal laser scanning microscope but was not successful. The Nikon MM-60 was used to measure the maximum depth of the indentations. Maximum indentation depths are shown in figure 6.

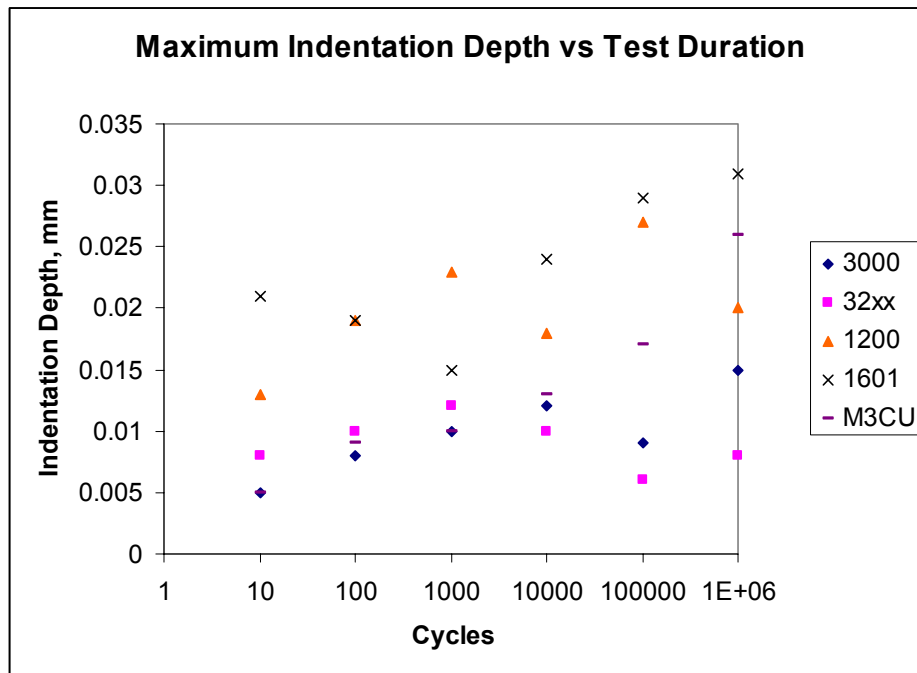


Figure 6: Maximum indentation depth vs test duration; 25mm diameter WC-6%Co ball impacting on five grades of PM steel.

Grazing incidence illumination was used in an attempt to locate the periphery of the indentations in order to measure their diameters, but this did not prove feasible. However the Nikon MM-60 was used to locate surface damage at indentation peripheries. This met with success in the case of the M3CU grade of PM steel; cracks were found which were associated with the presence of manganese sulphide (MnS) particles at several different test durations. An example is shown in figure 7.

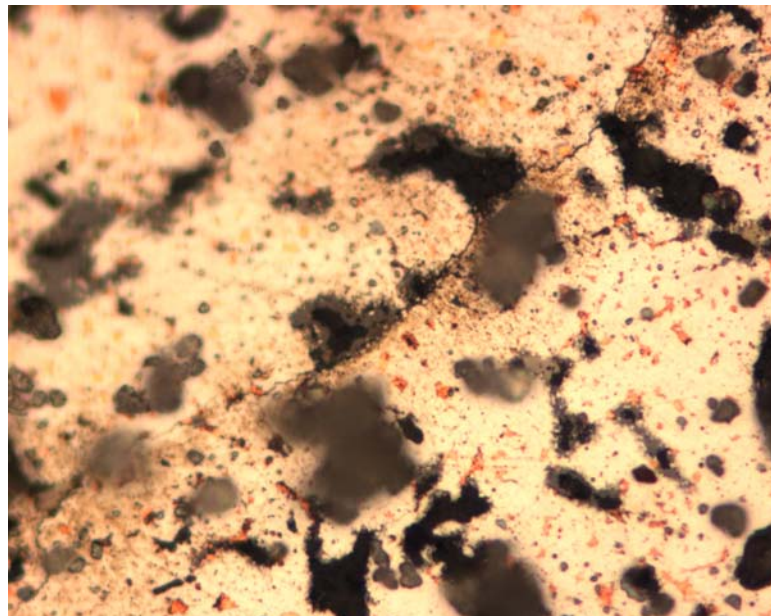


Figure 7: Damage in the form of a partial ring crack in M3CU subjected to 10^5 cycles at an indentation periphery. Crack runs through high speed steel matrix, carbides, copper and MnS particles.

Following the successful detection of surface damage on the polished and fatigued surfaces of M3CU, this grade was selected for further examination. The M3CU sample was sectioned vertically through the indentations, the sections mounted, ground and metallographically polished on diamond laps and polishing cloths. The final polishing stage was a relief polishing using a 0.05 μm colloidal silica suspension. Samples were then examined for subsurface damage using a Hitachi S-4000 field emission scanning electron microscope (FESEM) in direct electron mode (SEI). Subsurface damage was indeed found after the M3CU sample had been subject to various durations of fatigue test. In most cases the damage was associated with MnS particles. Examples are shown in figures 8 to 13.

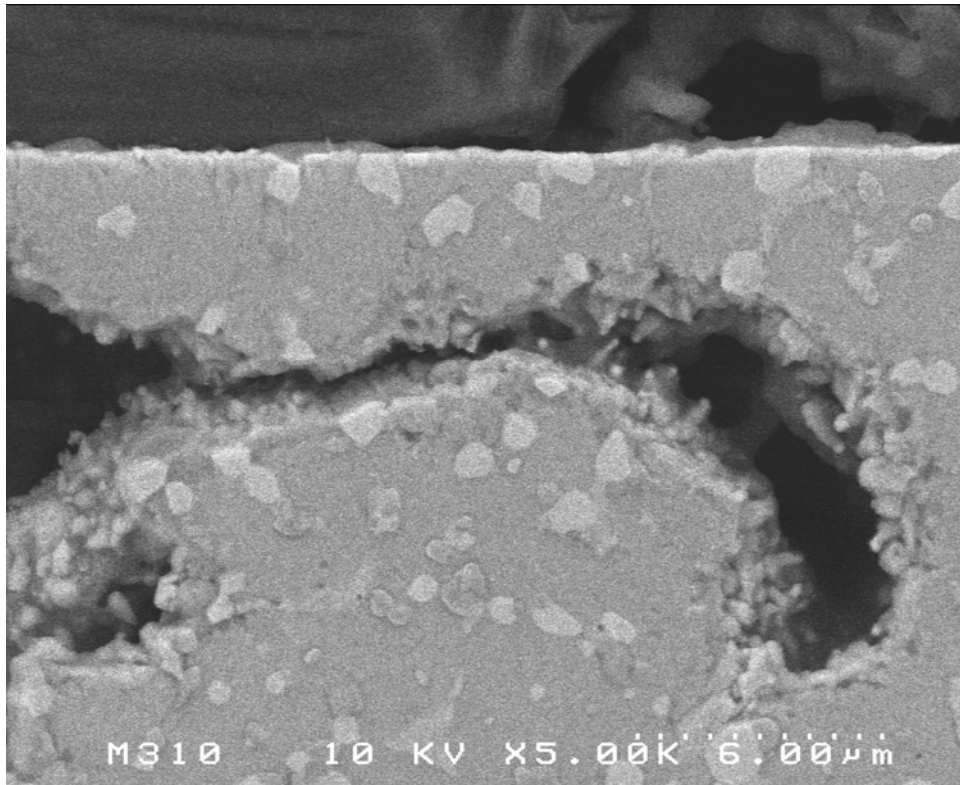


Figure 8: Possible subsurface damage in M3CU after 10 cycles.

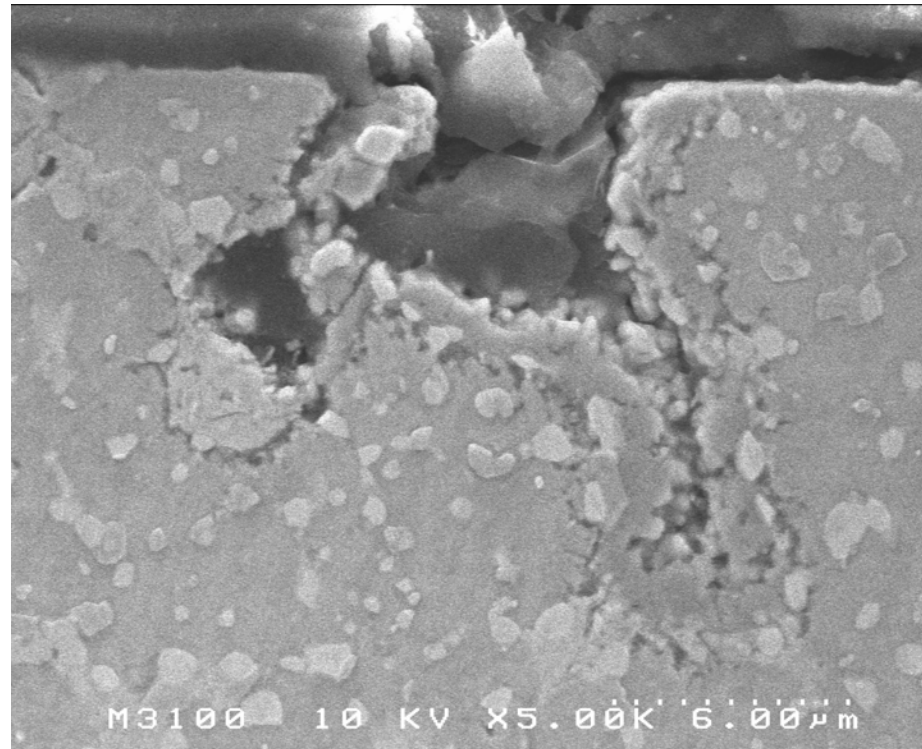


Figure 9: Subsurface damage in M3CU after 100 cycles; fracture/delamination associated with probable site of MnS particle.

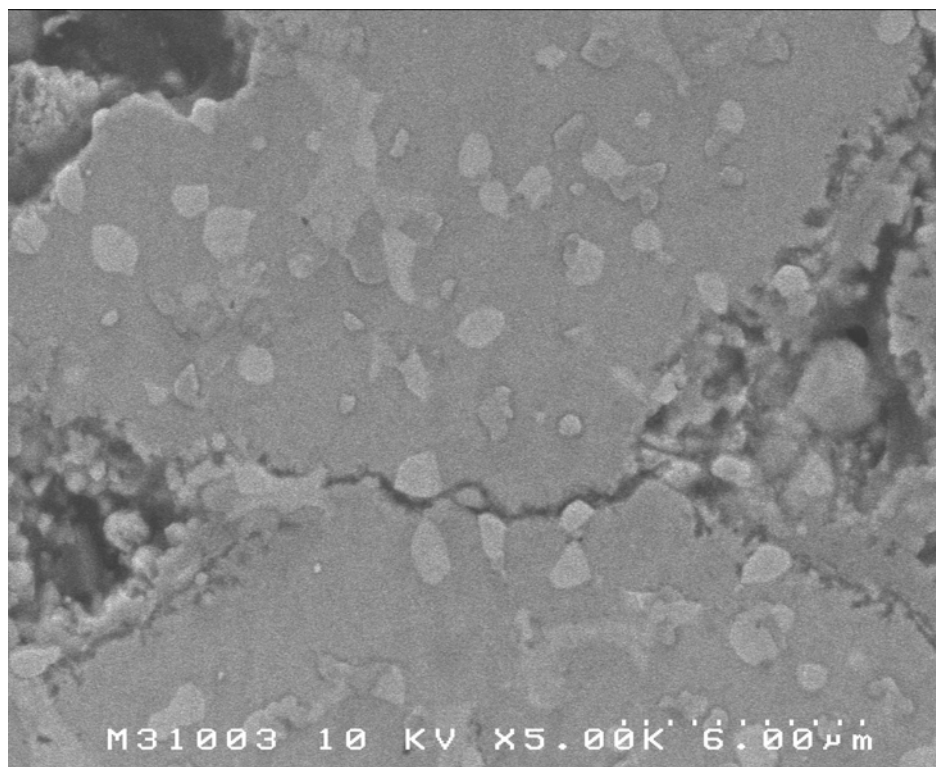


Figure 10: Subsurface damage in M3CU after 1000 cycles; fracture of high speed steel matrix with debonding of carbides between two MnS/copper regions.

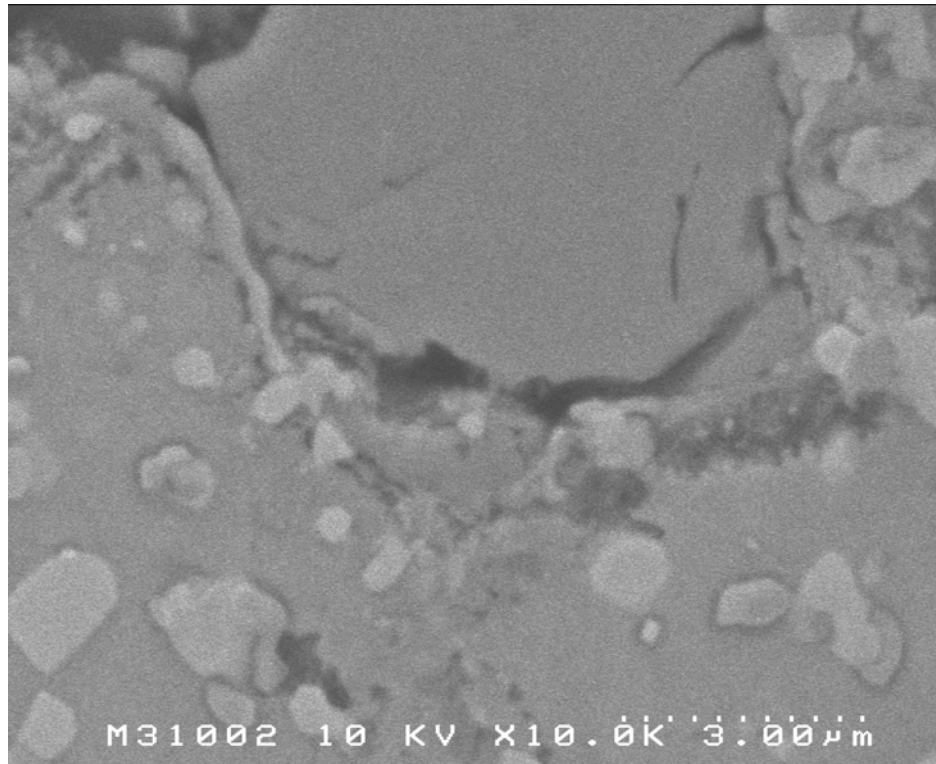


Figure 11: Subsurface damage in M3CU after 1000 cycles; fracture of high speed steel matrix.

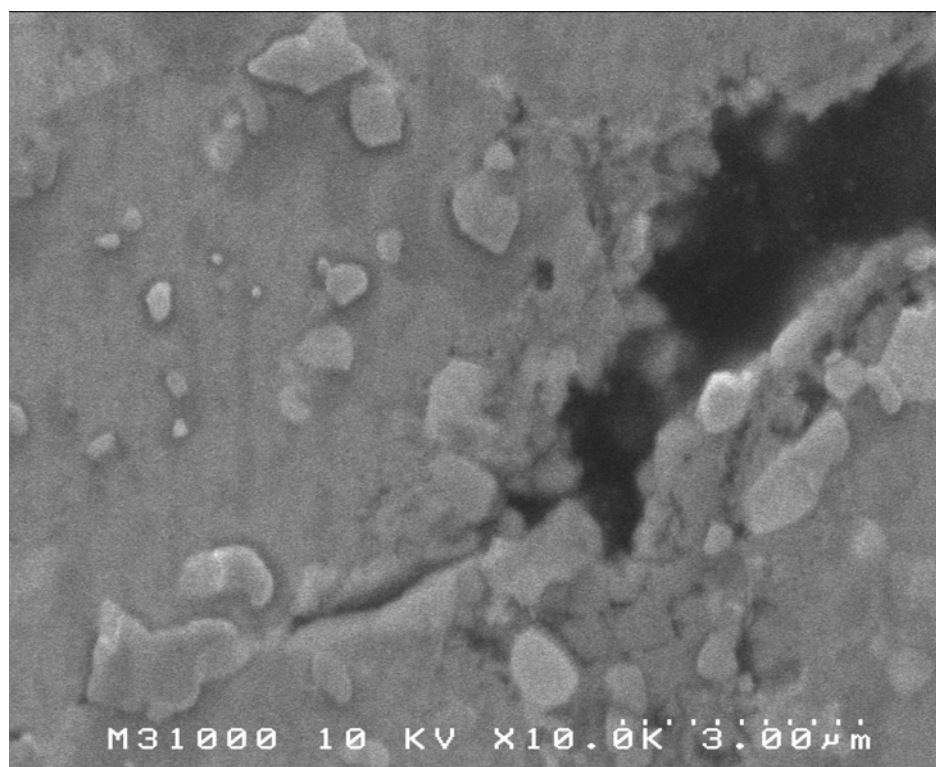


Figure 12: Subsurface damage in M3CU after 1000 cycles; fracture of high speed steel matrix; possibly initiated at an MnS particle.

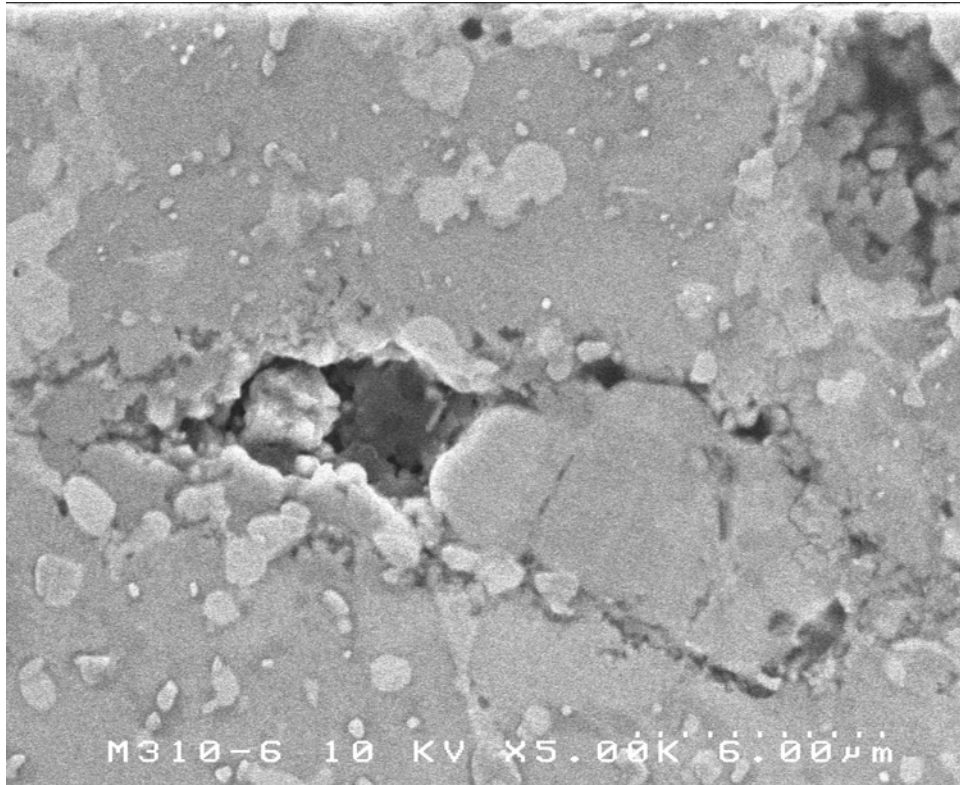


Figure 13: Subsurface damage in M3CU after 10^6 cycles; fracture of high speed steel matrix/debonding of second phase; possibly copper.

8 DISCUSSION

Five grades of powder metallurgy steel supplied by Federal Mogul Sintered Products Ltd have been subjected to dynamic loading using tungsten carbide based hardmetal balls. Initial static and quasi-static tests were conducted both at Federal Mogul and NPL. One positive outcome of this was close agreement between actual measured compressive yield strengths and those computed from hardness values using Meyer's law. However, using elastic-plastic contact mechanics it was found that predicted values for the initiation of plastic deformation could not be reconciled with those detected using an Instron test rig in compression. It may be the case that local yielding initiated at much lower values than predicted due to local microstructural effects giving rise to localised plasticity such as manganese sulphide particles, pores or copper. Damage in most samples after undergoing dynamic loading appeared to be confined to plastic deformation. However, one grade supplied; M3CU; did show evidence of partial ring cracks and subsurface cracking. Despite this there was not enough evidence to show whether this cracking was a form of fatigue response. M3CU was also the only grade supplied which showed gradual deepening of the indentations with increasing test duration.

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