Effect of SiC addition on mechanical and wear characteristics of WC-32(W-Ti)C-6Co cemented carbides

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Abstract: The effect of silicon carbide (SiC) addition on mechanical and wear properties of cemented carbides was investigated in this study. Different percentages (1–7 wt. %) of SiC were added to the cemented carbides mixture of WC/ 32 wt. % (Ti-W) C/ 6 wt. % Co. The microstructural characteristics of the developed materials was identified using scanning electron microscopy (SEM) and Rockwell-A macrohardness (HRA), Vickers microhardness (Hv) as well as transverse rupture strength (TRS) were measured. The experimental results revealed that inserting SiC inclusion into the cemented carbides is found to be useless for two reasons. First, the microstructure of the developed carbides has more aggregations of largely contiguous SiC grains led to presence of rich/poor regions and consequently poor compatibility between carbides and the binder. Second, all properties of the cemented carbides greatly declined with the addition of SiC particles.

Keywords: microstructure; SiC; mechanical properties; wear resistance; cemented carbides.

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1 Introduction

Cemented carbides are liquid phase sintered materials whose range of very hard, refractory and wear-resistant alloys (Brookes, 1996). The properties of cemented carbides are highly affected by their microstructural characteristics those rely on the chemical composition of the carbides, the matrix phases and on shape, size and distribution of the carbide particles. They are also affected by degree of the intersolubility of carbides (Upadhyaya, 2002; Angelo and Subramanian, 2008). Cemented carbides have superior hardness and toughness as well as good yield strength, abrasion and thermal stability. Therefore, they use in rigid industrial applications such as metal forming, rock drilling, mining tools and cutting tools.

The cemented carbide cutting tools have taken a large attention from both academic and industrial researchers to promote their properties aiming to allow faster machining, withstand higher strengths, and to leave a better finish on the part. Introducing a small amount (0.5–3 wt. %) of hard materials such as tantalum, niobium, chromium, vanadium, titanium, or hafnium to tungsten carbide/cobalt (WC/Co) carbides leads to modify their chemical composition, control grain growth, and enhance their deformation and high-temperature resistance (Brookes, 1996; Shin and Lee, 2006; Cha et al., 2001; Trent and Wright, 2000).

Numerous studies have been done to investigate the effect of introducing inorganic compounds on microstructure, mechanical behaviour and cutting performance of the cemented carbides. For example, Guo et al. (2008) investigated the effect of adding Mo_2C to the WC-NbC-Co carbides, and it was found that the wettability between carbides and the binder improved and a core/rim structure formed with the existence of Mo_2C carbide. Huang et al. (2008) showed that doping WC-TiC-Ni by VC and Cr_3C_2 carbides inhibits the grains growth of the WC and improves the mechanical properties of the carbides. Saito et al. (2006) investigated the effect of WC grain size and Co content on microstructure and wear resistance of the WC carbide. The wear properties increased with increasing both WC particle size and Co content while the change in the carbide's grain size resulted in presence of porosity in the final microstructure.

Furthermore, Correa et al. (2010) swapped the Co binder in WC-Co cemented carbides by Ni and Si binder. Vickers hardness and flexural strength values of the developed WC-Ni-Si were found to be similar for those of the WC-Co carbides. Wei et al. (2012) studied the role of carbon (C) addition on microstructure and properties of the WC-Co cemented carbides. Obtaining excellent properties of the WC-Co carbides with a pure phase constitution depends on the WC grain structure and a suitable carbon addition in the final sintered bulk. The amount of η phase (Co₆W₆C) in the final sintered bulk declines with increasing carbon addition up to 17 wt. % or further.

On the other hand, ceramic materials such as silicon carbides (SiC) is commonly made cermet cutting tools (ceramic (cer) and metallic (met)) for their high-temperature strength, hardness and they have excellent wear and corrosion resistance (Kelly and Zwebin, 2000; Liversage et al., 2007). Moreover, SiC is efficiently used as grinding wheels through the use of glues known as bonds that do not confer atomic contact; they provide porosity for cutting fluids to enter them and also for miniscular grinding chips to get lodged inside them (Venkatesh and Chandrasekaran, 1987). The potential of incorporating low proportions of silicon carbide (SiC) into cemented carbides to produce newly developed composites at relatively low cost would be not a well-adjusted process not only with regards to their properties but also its sintering behaviour. This is because

the difference in solubility of ceramics and cemented carbides by the binder. Cermet tools are often manufactured with nickel rather than cobalt binders, so that the reactions involved in chemical and oxidation wear differ from those encountered with WC tools. However, Cermet tools typically fail by mechanisms similar to those for WC tools, which is, by abrasion, plastic deformation, edge chipping and depth of cut notching due to oxidation (Stephenson and Agapiou, 2006). Hence, the current study is made to find out the role of merging SiC particles with WC-32(Ti-W)C-6Co cemented carbides on microstructure, mechanical properties and wear resistance of the produced carbides, and to determine the advantages or disadvantages of using SiC with cemented carbides.

2 Experimental

2.1 Materials and preparation method

Cemented carbides of WC, 32 (Ti-W)C and 6 Co wt. % supplied by (HCST Berlin GmbH, Goslar, Germany) with average sizes of 2.5, 2.5 and 1.35 µm, respectively, were blended in 4.5 dm³ rotary ball milling. Acetone was used as mixing aid at ratio of 1 : 4 for the total carbides weight and the blending process was lasted for 12 h. The slurry was then dried in a vacuum furnace at 90°C to remove the acetone and the average grain size of the mixture was Fischer sub-sieve size (FSSS) of 5.78 µm after passing through sieve with 30 meshes to remove the solid balls. Paraffin wax was added as a die lubricant at 2 wt. %. Different weight fractions, 1, 3, 5 and 7% of SiC powder with an average grain size of 4.26 µm and an apparent density of 0.92 g/cm³ (bought from ALM-41-sumito, UK) were then added to the carbide mixture and remixed by a high shear mixer to avoid a prospective segregation. The cutting tips of (TNMG-220416) shape, ISO standard, 60° rhomboid type (Figure 1) were made by powder metallurgy technique at 170 MPa double die compaction pressure and sintered at 1410°C for 1 h soaking in a vacuum furnace following the thermal process shown in Figure 2. The produced carbide alloys and their chemical compositions are listed in Table 1.

	Tool's name	Elements (wt. %)				Average grain	Apparent density
Alloy		SiC	WC	(Ti-W)C	Со	size (µm)	(g/cm^3)
А	A0	0	62	32	6	5.78	2.9
В	B1	1	61.38	31.68	5.94	6.32	2.83
С	C3	3	60.14	31.04	5.82	6.18	_
D	D5	5	58.9	30.4	5.7	6.27	_
Е	E7	7	57.66	29.76	5.58	6.33	-

 Table 1
 Chemical composition and some details of the carbides after adding SiC powder

2.2 Characteristics

Scanning electron microscope (SEM, JEOL-5900LV, Japan) was used to examine the microstructural behaviour of cemented carbides cutting inserts. For this purpose, specimens with $5 \times 5 \times 5$ mm³ were cut from carbides by electro-discharge machining

(EDM) technique. The specimens were polished by different grinding and lapping discs before they were sputter coated by gold to improve their conductivity.



Figure 1 Schematic graph of the cemented carbide cutting inserts (see online version for colours)

Figure 2 The thermal cycle for the sintering process



Green and sintered densities of the carbides inserts were measured by Archimedes method. The shrinkage percent was calculated by the difference between specimen's dimensions before and after sintering process.

Rockwell-A (HRA) hardness was measured using 60 kg load, diamond indenter. Vickers hardness (Hv) was determined by using 0.98 N load and applying equation (1).

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$$Hv = 1.8544 \frac{P}{d^2}$$
(1)

where: P is the applied load and d is the average length of two diagonals of the indentation.

Transverse rupture strength (TRS) was calculated according to ASTM B 528 standard (Standard, 2010) using 5 tons applied load at crosshead of 0.5 mm/min and using equation (2). The TRS specimens of $5.2 \times 5.5 \times 51$ mm³ were prepared by following the same conditions of powder metallurgy process that were applied for preparing of carbides cutting tools.

$$TRS = \frac{3PL}{2wt^2}$$
(2)

where TRS is the transverse rupture strength (MPa), P is the force required to rupture the specimen (N), L is the span distance between the supported centres (25.4 mm), w is the specimen width (mm), and t is the specimen thickness (mm).

2.3 Wear resistance

In this study, the flank wear width (VB_B) was considered as a life criterion for assessing the suitability of carbide cutting tools. Lathe machine with a tool holder of $20 \times 25 \times 120 \text{ mm}^3$ with -11° rake angle, $+6^\circ$ flank angle and 60° relief angle was used for a flank wear evaluation. The VB_B of two kinds of carbide tools (*A0* and *C3* from Table 1) was measured using an optical microscopy. The VB_B measurements were carried out after machining of a medium carbon steel (CK-45) workpiece (the chemical composition of CK-45 is shown in Table 2) for one minute. Different parameters of machining process with no cooling fluid were used as shown in Table 3 (Albdiry and Almosawi, 2011). After each minute of continuous machining, the machine is aborted and the VB_B value is measured. The on/off machining processes last until the flank wear width reaches or just exceed 300 µm (0.3 mm) as a maximum failure gauge.

3 Results and discussion

3.1 Microstructure

The microstructure of sintered plain WC/32(Ti-W)C/6Co cemented carbides was inspected by Scanning Electron Microscopy (SEM), and shown in Figure 3. The dominant carbide WC grains and TiC grains in the microstructure of the plain cemented carbides (alloy-A) seems to be well-dense and well-dispersed in the metal Co binder. No apparent pores can be seen in the microstructure when compared to the newly developed cemented carbides with SiC grains (Figure 4(a)–(d)). The pore-free structure with excellent densification of WC or TiC grains during liquid phase sintering obtained due to a very good wetting of these carbides by the liquid metal binder, which consequently result in high strength, toughness and high hardness.

Table 2The chemical composition of steel work-piece used in this study

Workpiece	Materials (wt. %)						
code (DIN)	С	Si	Mn	Р	S	Fe	Vickers hardness (Hv)
Steel, CK-45	0.42	0.15	0.65	0.036	0.035	Bal.	220

 Table 3
 Lathe-machine cutting parameters used for wear assessment

Cutting speed (m/min)	Feed rate (mm/rev)	Depth of cut (mm)
68	0.14	0.5
68	0.48	0.5
146	0.14	0.5
146	0.48	0.5

On the other hand, the addition of SiC grains to the cemented carbides induced lower uniformity and distribution of carbides in the Co binder i.e., perceivable SiC agglomerates appeared in the structure even at small concentration of SiC (see Figure 4(a)). The core/rim structure appeared with introduction of SiC to the cemented carbides (see Figure 4(b) and (c)). The formation of the core/rim structure can be attributed again to the less interaction between SiC and the binder or SiC with other cemented carbides. Rich/poor regions of SiC accumulations also appeared in the metal binder (see Figure 4(d)), which may be due either to the decomposition of SiC grains into mullite matrix or finely-divided glass phases combining quartz and cristobalite (SiO₂) (Kelly and Zwebin, 2000). Or due to the absence of the encapsulation between SiC grains and the cemented carbides that should occur during the solid-state sintering stage. To obtain an optimum interface between carbides and other reinforcements in the cemented carbides, in essence the binder must wet these carbides or reinforcements (Christensen and Wahnstrom, 2006).

The addition of rigid components to cemented carbides drastically hinders the interaction between adjacent particles and the binder and forms grain boundaries. The grooves of the grain boundaries along with an irregular shape of SiC grains i.e., triangular-like shape that induced unevenly carbides distribution (see again Figure 4(d)), caused in poor interlocking, low level of densification and act as stress raisers (Angelo and Subramanian, 2008; Aigbodion and Hassan, 2007). In order to improve the properties of the cemented carbides and improve their morphological behaviour with the use of a newly added component, it is necessary to understand the affinity or the interfacial bonding between the additive and the binder. And understanding the matching between the additive and other carbides (Liang, 2002; Wei et al., 2012). In contrast, poor interfacial bonding between the additives and the binder may result in particles' debonding and extensive localised plastic deformation in the final product.

3.2 Characteristics

Figure 5 shows the relationship between green density, sintered density and shrinkage of WC-32(Ti-WC)C-6 Co carbides vs. different SiC percentages. Both densities of the newly produced cemented carbides decreased with increasing SiC values. This decrease is obviously because SiC powder has a very low apparent density of 0.92 g/cm³

compared to 2.9 g/cm^3 for the cemented carbides. Also, the sintering stages and temperature that they are more suitable to cure cemented carbides but would not capable to determine full densification of cemented carbides doped by ceramic carbides (Brevala et al., 2005).

Figure 3 SEM image of the alloy A pressed at 170 MPa and sintered at 1410°C under vacuum



Figure 4 SEM micrographs of obtained cemented carbides with different SiC additions: (a) alloy B; (b) alloy C; (c) alloy D and (d) alloy E (see online version for colours)



Figure 5 also shows an increase trendline of shrinkage values with increasing SiC content. This is due to that a solid-state shrinkage is highly dependent on uniform distribution of particles in a liquid-state sintering stage. And since the microstructure is unevenly distributed, the shrinkage is obviously increased.

Rockwell macrohardness (HRA), Vickers hardness (Hv) and Transverse Rupture Strength (TRS) of sintered carbides are shown in Figure 6. Modest increase in the HRA occurred with introducing SiC up to 3 wt. %. This is ascribed to the reduction in decarburisation (brittle η phase) with introducing SiC. The reduction in the η phase occurs due to a higher carbon content (of 30% C) in the chemical composition of SiC compared to tungsten carbide (Kelly and Zwebin, 2000). However, the macrohardness declined from 83 HRA (at 3% SiC) to 68 HRA for carbides contain 7% SiC. This significant reduction in HRA can be attributed to presence of binary brittle compounds i.e., Tungsten silicide (WSi₂) or Titanium silicide (TiSi₂) in the solid state of the cemented carbides. Using SEM analysis would not be capable to examine η phase or even the other phases that occur in the final microstructure of the carbides. The positron annihilation lifetime spectroscopy (PALS) may be more suitable to determine type and volume of each phases occurred in the sintered carbide's structure.

From Figure 6, it can also be shown that Vickers hardness (Hv) of the cemented carbides greatly decreased with increasing SiC. The Hv of alloy A (0% SiC) was 1200, while it dropped to 1033 at 1% SiC and 848 (30% reduction) and 667 (45% reduction) at 5% and 7% SiC, respectively. This is obviously attributed to a poor dispersion of SiC in the Co binder and a poor interlocking between SiC grains and the cemented carbides.



Figure 5 Physical properties of cemented carbides cutting inserts vs. different SiC additions compacted at 170 MPa and sintered at 1410°C under vacuum (see online version for colours)



Figure 6 Mechanical properties of carbides' cutting tools vs. different SiC contents pressed at 170 MPa and vacuum sintered at 1410°C (see online version for colours)

Transverse rupture strength (TRS) of cemented carbides significantly reduced with introducing SiC into them (see Figure 6). The TRS of alloy-A was 1940 MPa and it declined to 1720 MPa with a 1 wt. % SiC addition. This reduction is due to the deficiency of a soft liquid binder after introducing SiC. Also, the unevenness distribution of the carbide phases in the binder and the subsequent porosity thereof can be another reason of crack initiation, crack propagation and consequently a rapid fracture of the part. Since the heterogeneous distribution of the carbides in the microstructure can act as stress raisers or precursor of a brittle fracture (Ray et al., 2008; Lengauer and Danzer, 2008; Albdiry and Alethari, 2012). Similarly, Wei et al. (2012) showed that the addition of higher contents of carbon to the carbides caused in crack nucleation and propagation as well as low strength of the final part.

3.3 Wear characteristic

The flank wear measurements of carbide cutting tools are shown in Figure 7. In general, the wear resistance of both kinds (A0 and C3) of carbide tools decreases with increasing cutting speed due to the generated heat at the interface contact between the tool and the workpiece with high cutting speeds. The high temperature on the tool's surface would soften the tools and induces low shear strength with a localised plastic deformation and a bulge in the tools. The A0 tool's life was longer than that of the C3 tool by minute i.e., the former tool can efficiently machine the workpiece higher than the latter tool by minute before reaching 0.3 mm. This is because the incomplete welding between the carbides' particles and the binder that adversely causes in low densification, low hardness and consequently low wear resistance of the cemented carbides (Albdiry and Almosawi,

2011; Shaw, 2005). This phenomenon is perhaps due to the insufficient compaction pressure of 170 MPa since the ceramics might require higher pressures and higher sintering temperatures compared to the cemented carbides.

As a brief description of the machined tools' lives in numbers, the A0 tool efficiently lasted up to 5.25 minute as a machining time of the workpiece at 68 m/min cutting speed, 0.14 mm/rev feed rate and 0.5 cutting depth, before reaching the flank wear width of 0.3 mm. Whereas the C3 tool was successfully operated up to 3.5 minute using the identical cutting conditions for the A0 tools. Similarly, at higher cutting conditions of 146 m/min speed and 0.48 mm/rev feed rate, the machined time of the A0 tool was more than twice the time of the C3 tool at 130 and 53 seconds, respectively.



Figure 7 The flank wear widths of two kinds of the cemented carbides, the A0 and C3 tools vs. cutting machining time (see online version for colours)

4 Conclusions

The current paper is done to investigate the role of SiC addition on microstructure, mechanical and wear characteristics of WC-32(Ti-W)C-6Co cemented carbides. The main findings of the work are: the microstructure of the chosen cemented carbides is highly susceptible with introducing foreign inclusions such as SiC powder even at a small amount. SiC in the sintered carbides does not interact with the WC powder at the final consolidation. So that SiC has significantly reduced the properties (hardness, density and

rupture strength) of the cemented carbides. The reduction in these properties was attributed to poor affinity or not existed at all from SiC to WC that result in segregation and act stress concentration. The wear resistance of the carbides' cutting tools also decreased with introducing SiC powder to the plain cemented carbides due to the presence of SiC agglomerates in the sintered part.

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