United States Patent [19]

Rudy

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[45] **Dec. 18, 1973**

[54] **CARBIDE ALLOYS SUITABLE FOR CUTTING TOOLS AND WEAR PARTS**

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Related U.S. Application Data

[63] Continuation-in-part of Ser. No. 802,625, Feb. 26, 1969, Pat. No. 3,690,962.

[52] **U.S. Cl.** ..... **75/176,** 75/134 F, 75/134 V, 75/175.5, 75/177

[51] **Int. Cl.** **C22c 27/00,** C22c 29/00

[58] **Field of Search** ..... 75/176, 134 F, 134 V, 75/175.5, 177

[56] **References Cited**

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[57] **ABSTRACT**

This invention relates to refractory metal bonded carbide alloys for use as cutting tools and in other applications where high hardness and abrasion resistance are required. The desired fine-grain, lamellar microstructure is obtained preferably by casting eutectic, or near-eutectic composition alloys of a Group IVa metal (titanium, zirconium, hafnium), tungsten and carbon which may contain certain alloying and inert materials. For selected applications, the composites may be fabricated by powder-metallurgical techniques.

5 Claims, 5 Drawing Figures

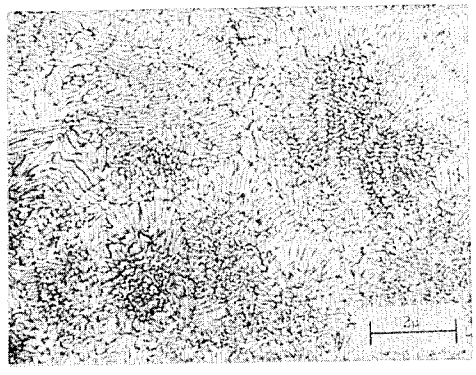


FIG. 1.



FIG. 2.

Fig. 3

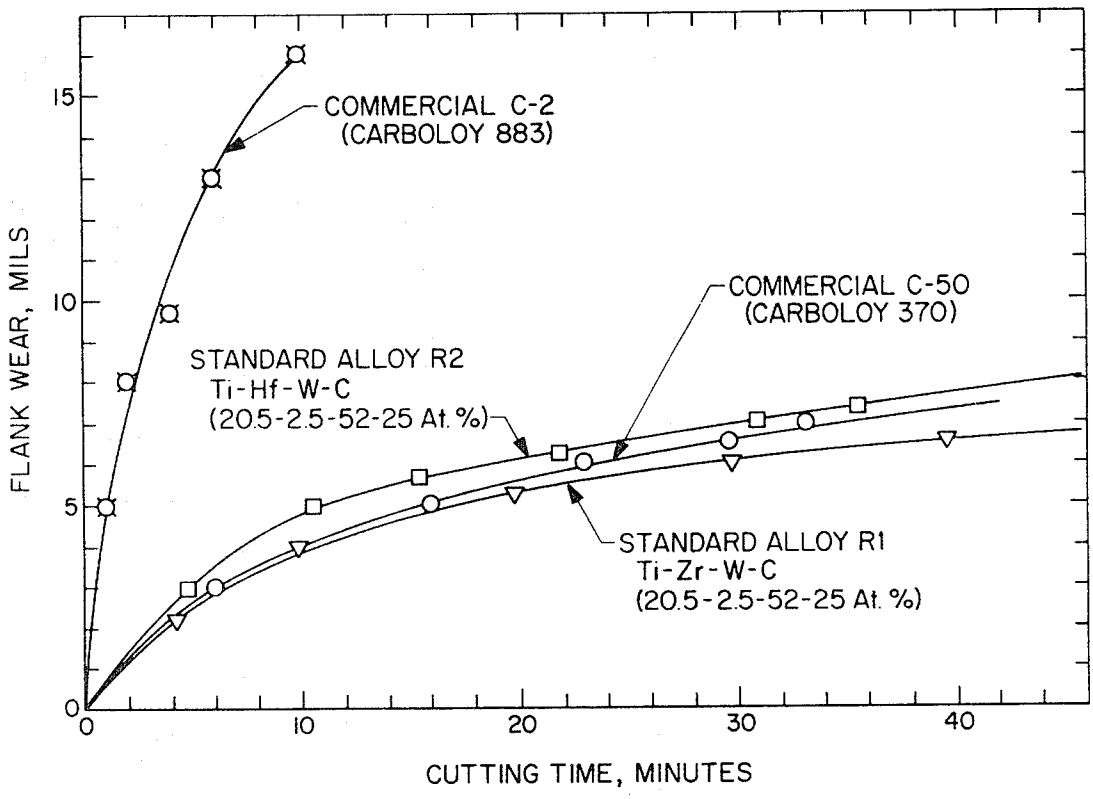

FLANK WEAR, MILS
0
5
10
15
0
10
20
30
40
CUTTING TIME, MINUTES
COMMERCIAL C-2
(CARBOLOY 883)
COMMERCIAL C-50
(CARBOLOY 370)
STANDARD ALLOY R2
Ti-Hf-W-C
(20.5-2.5-52-25 At.%)
STANDARD ALLOY R1
Ti-Zr-W-C
(20.5-2.5-52-25 At.%)

Fig. 4

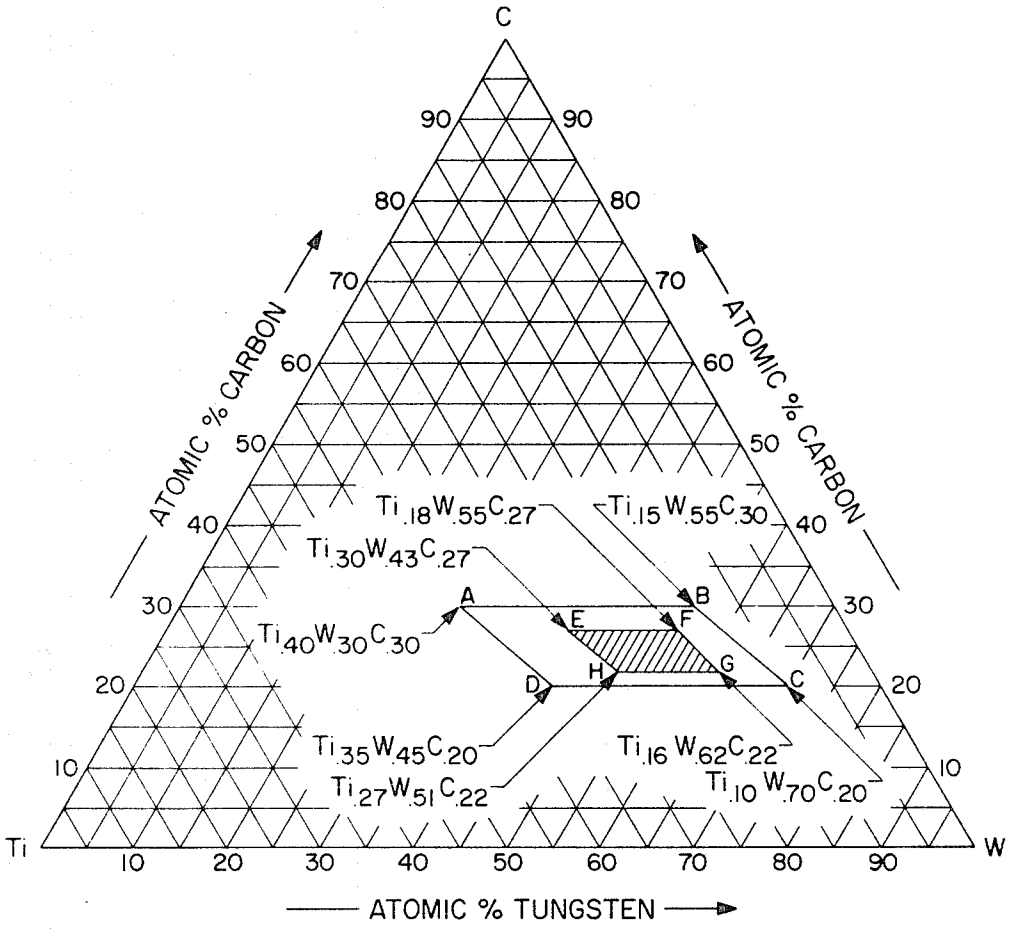
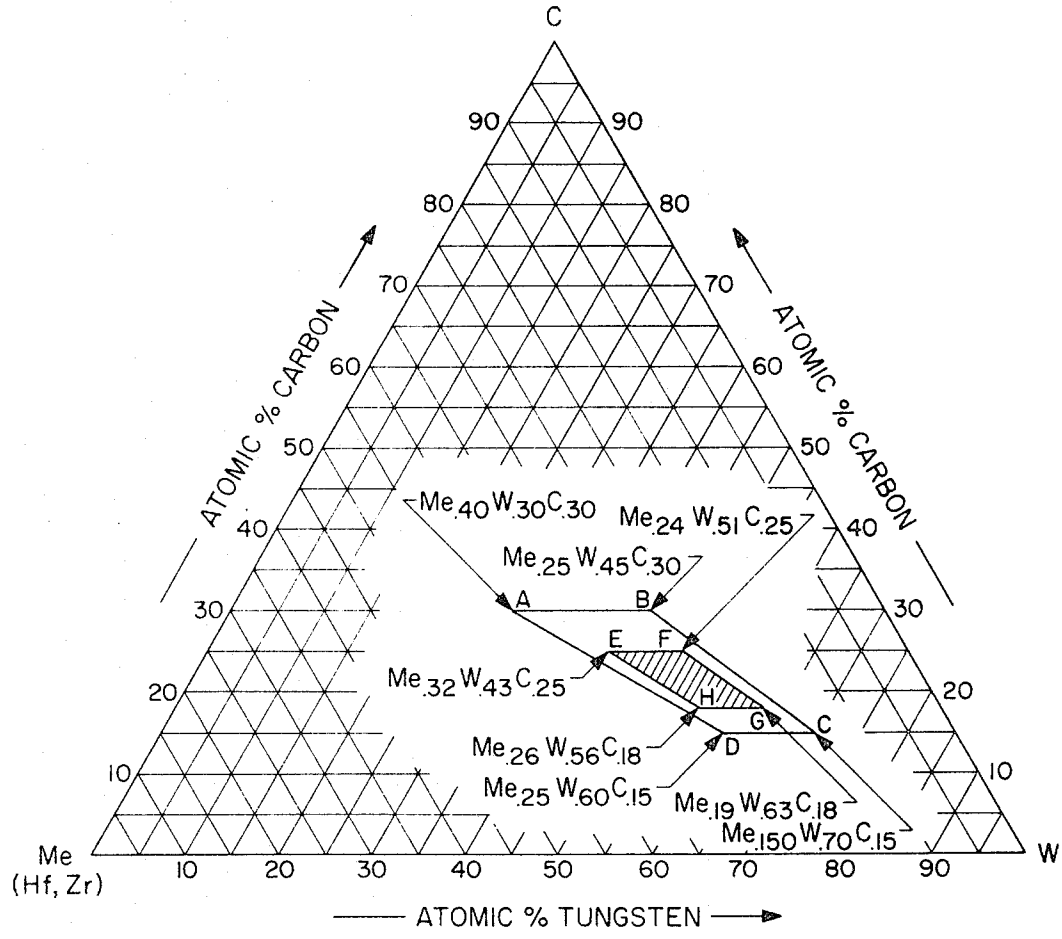

C
90
80
70
60
50
40
30
20
10
ATOMIC % CARBON
ATOMIC % CARBON
Ti
W
10
20
30
40
50
60
70
80
90
ATOMIC % TUNGSTEN
Ti.18W.55C.27
Ti.15W.55C.30
Ti.30W.43C.27
Ti.40W.30C.30
Ti.35W.45C.20
Ti.27W.51C.22
Ti.16W.62C.22
Ti.10W.70C.20
A
B
C
D
E
F
G
H

Fig. 5


C
90
80
70
60
50
40
30
20
10
ATOMIC % CARBON
ATOMIC % CARBON
Me.40W.30C.30
Me.24 W.51 C.25
Me.25 W.45C.30
A
B
E
F
Me.32 W.43 C.25
H
G
C
Me.26 W.56C.18
D
Me.25 W.60C.15
Me.19 W.63C.18
Me.150W.70C.15
Me
(Hf, Zr)
10
20
30
40
50
60
70
80
90
W
ATOMIC % TUNGSTEN

CARBIDE ALLOYS SUITABLE FOR CUTTING TOOLS AND WEAR PARTS

This application is a continuation-in-part of copending application of Erwin Rudy Ser. No. 802,625, filed Feb. 26, 1969, Pat. No. 3,690,962, issued Sept. 12, 1972.

DISCUSSION OF THE PRIOR ART

Modern carbide tooling materials consists of a mechanically-pulverized, hard carbide phase dispersed in a matrix (binder) of an iron group metal, usually cobalt or nickel. The binder phase contributes toughness to the composite and also serves as an aid in sintering the carbide particles. The loss of strength of iron metal-based binder phases at relatively low temperatures can cause thermal wear to become the dominant wear mechanism at high cutting speeds. The low melting temperatures of these binder phases also preclude their use as abrasion-resistant composites at temperature above 800°C. to 1,000°C.

Binderless, cast carbides such as W_2C + WC eutectics played a role in the initial development of carbide-based tools and die materials, but became obsolete with the advent of the tougher, cobalt-bonded carbides fabricated by powder-metallurgical techniques.

Despite the attractive features of the casting process, including its adaptability to low cost manufacturing methods and need for only moderate capitalization, castable tooling materials of equivalent performance to the iron group metal bonded carbides were not developed.

DESCRIPTION OF THE INVENTION

The carbide composite materials of this invention have excellent thermal and mechanical shock resistance compared to the conventional cobalt-bonded carbide tool. This is achieved through the formation of a fine-grain, lamellar microstructure having a hard monocarbide phase and a tough refractory metal phase.

The carbide composite of this invention, in a preferred embodiment, comprises a base alloy system of a Group IV*a* metal (Ti, Zr or Hf), tungsten and carbon having a fine-grain, lamellar microstructure which is derived from a pseudobinary eutectic or near-eutectic composition. The lamellar microstructure possesses a monocarbide cutting phase and a metal phase with the latter phase being rich in tungsten and contributing toughness to the composite. The monocarbide phase contains significant amounts of both the tungsten and the Group IV*a* metal. In a preferred embodiment (hypereutectic composition) of the carbide composite of the invention, grains of primary carbide are dispersed throughout the lamellar microstructure. The interspersed primary carbide grains significantly improve the cutting action of the composite when employed as a machining tool.

The carbide alloys of the invention are made possible by the existence of a pesudobinary eutectic (see Table I) in the systems of Ti—W—C, Zr—W—C, and Hf—W—C. The solidification of the eutectic liquid provides a refractory metal phase and a monocarbide phase that are in solid state two-phase equilibria. The co-existence of the metal phase and the monocarbide phase in the solid state microstructure is necessary to the concept of the metal-bonded carbide tool of the invention. The occurrence of eutectic or near-eutectic-composition reactions provides the means by which the desired microstructure, which possesses an extremely fine-grain, lamellar mixture of metal and monocarbide phases, can be obtained by melting and casting.

Table I shows the exact eutectic compositions and the compositions of the coexisting component phases of each of the three respective eutectics in the Ti—W—C, Zr—W—C, and the Hf—W—C systems.

TABLE I.—DATA FOR THE PSEUDOBINARY EUTECTIC EQUILIBRIA IN THE SYSTEMS OF Ti-W-C, Zr-W-C, AND Hf-W-C

		Compositions (In atomic percent)		
Alloy system	Eutectic temperature (°C)	Eutectic liquid	Metal—solid solution (approximate)	Monocarbide—solid solution (approximate)
Ti-W-C	2700	Ti(21)-W(57)-C(22)*	Ti(8)-W(92)-C(< 0.5)	Ti(28)-W(33)-C(39)
Zr-W-C	2850	Zr(25)-W(55)-C(20)*	Zr(3)-W(97)-C(< 0.5)	Zr(44)-W(16)-C(40)
Hf-W-C	2980	Hf(21)-W(61)-C(18)*	Hf(4)-W(96)-C(< 1)	Hf(40)-W(20)-C(40)

* Carbon concentrations ± 1 atomic percent, metal concentrations ± 2 atomic percent.

Alloy compositions from the Ti—W—C systems, even when they are located somewhat away from the eutectic composition, will still solidify almost isothermally and thus produce acceptable micro-structures closely resembling those of the true eutectics. Hence, there is in this alloy system, considerable latitude in varying the properties of the composites by changing the overall composition of the alloys without losing castability or changing the basic morphology of the cast structure derived from the pseudobinary eutectic or near-eutectic reaction.

There is considerably less freedom in varying the compositions of the Zr—W—C and Hf—W—C alloy systems, and in both of the latter systems it has been found that solidification occurs over a relatively wide temperature range when the compositions are located more than approximately **10** atomic percent to the zirconium or hafnium side away from the pseudobinary eutectic. Hence, the microstructure of the latter two systems, if provided with too much zirconium or hafnium, will have a coarser grain structure and the composites will be less suitable for cutting tool applications.

Other advantages of the invention will be apparent from the following detailed descriptions and drawings in which:

FIG. **1** is a photomicrograph of a typical Group IV*a* metal (in this instance titanium)-tungsten-carbon composition system which contains (Ti(21)—W(5-7)—C(22) in atomic percents) taken at a magnification of 1000X;

FIG. **2** is a photomicrograph at a magnification of 500X of another Group IV*a* metal-tungsten-carbon system of somewhat different composition namely (Ti(2-

3)—W(52)—C(25) atomic percent) wherein there are grains of primary carbide dispersed throughout the lamellar microstructure;

FIG. 3 is a graph presenting typical comparative wear curves obtained in turning Type 347 stainless steel with the cast alloy tools prepared according to the invention and with top grade C—2 and C—50 type commercial carbide tools.

FIG. 4 is a compositional ternary diagram showing desired composition areas for Ti—W—C base alloys of the invention; and

FIG. 5 is a compositional ternary diagram showing desired composition areas for Zr—W—C and Hf—W—C base alloys of the invention.

The carbide composites of the invention are preferably prepared by melting and casting to produce the fine-grain, lamellar micro-structure of monocarbide phase and refractory metal phase formed through solidification of an eutectic or near-eutectic-composition liquid. A typical fine-grain lamellar microstructure of the invention is illustrated in FIG. 1 where the metal is dark and the carbide light. The photomicrograph of that figure, while showing a titanium-tungsten-carbon system, is typical of the lamellar microstructure of all three base alloy systems, i.e., Ti—W—C, Hf—W—C, and Zr—W—C of the invention.

For carbide cutting tool applications it has been found desirable to have grains of primary carbide dispersed throughout the lamellar microstructure as seen in FIG. 2. The photomicrographs of the hafnium and zirconium systems show similr microstructures to those of FIGS. 1 and 2. The presence of the grains of primary carbide in the lamellar structure significantly improve the use of the carbide composite for machine tool cutting purposes.

The ternary diagrams of FIGS. 4 and 5 depict base alloy compositions suitable for producing the carbide composites of the invention. Referring to FIG. 4, which is concerned with titanium-tungsten-carbon base alloys, it is seen that the preferred compositions fall within the inner hatched area E, F, G, H. The larger area A, B, C, D includes compositions of generally less suitable composites, but which are acceptable for some applications. Similarly, the inner hatched area of E, F, G, H of FIG. 5 dpicts the more desirable compositions of either the hafnium or zirconium systems. The larger enclosed area A, B, C, D includes composites generally less suitable, but still of a useful nature. The lamellar microstructures of the preferred areas E, F, G, H of FIGS. 4 and 5 include grains of primary carbide dispersed throughout the microstructure. These primary carbide grains enhance the cutting characteristics of the composites when used in machine tools. However, too much of the primary carbide grains promotes chipping of the machine tool. Compositions falling within the general areas A, B, C, D graphically above the preferred areas E, F, G, H will have some tendency to chip. Such is tolerable for some machine tool applications and not objectionable at all for other applications where high hardness and abrasive resistance are required. Below the preferred areas E, F, G, H of FIGS. 4 and 5 but within the area A, B, C, D there is a tendency for primary metal grains to form within the lamellar micro-structure. Primary metal lessens the value of the carbide composite when used as cutting tools. To the right of the preferred areas E, F, G, H of both FIGS. 4 and 5 and within the general areas A, B, C, D there is a tendency for subcarbide grains to form within the lamellar micro-structure. Subcarbides are less hard than monocarbides and, therefore, less suitable for machining applications. The inner hatched areas E, F, G, H of FIGS. 4 and 5 contain the primary carbide grains in desired amounts for most machine tool purposes.

Melting and casting, plasma-arc spraying, as well as powder-metallurgical methods have been employed in preparing metal and monocarbide composites based on the alloy systems Ti—W—C, Zr—W—C, and Hf—W—C of the invention. Melting followed by casting into chilled molds has produced composites with the best mechanical properties and performance for cutting tool purposes. Experience indicates that skull melting, a technique using either a nonconsumable (tungsten) or a consumable electrode, is the most efficient and reliable method for obtaining the melts needed for casting. Melting of the charges in resistively or inductively heated graphite containers has been proven feasible for Ti—W—C base alloys, although care has to be exercised to avoid excessive carbon pick-up upon prolonged exposure of the alloys to hypereutectic temperatures. Continuous melting of presinerted compacts in the field of an eddy-current concentrator, or resistance heating and melting of alloy charges in an arrangement where the container is formed by a solidified portion of the alloy to be melted, appear to be promising techniques.

Centrifugal casting of the melt is preferable to casting techniques employing stationary molds, because the former casting techniques minimize the problems associated with the formation of shrinkage pipes and, as the result of the high casting speeds, allows complex parts to be cast to shape.

Other uses of the alloys of the invention are many including hard facings for plows, bulldozer blades, bearings, and for penetrator cores for armor-piercing projectiles. Because of their relatively low neutron capture cross section, alloys containing zirconium, and based on the Zr—W—C combination, are thought to find possible use in nuclear applications. In ceratin instances where small amounts of Titanium, Hafnium, or both, in quarternary or higher order combinations such as (Zr, Ti)—W—C, (Zr, Hf)—W—C or (Zr,Hf,Ti)—W—C are tolerable, multi-Group IV*a*-metal containing alloys may be more economical, or may be utilized when an increase in hardness is desired in spite of the increased neutron capture cross section.

Application of hard facings to various shaped objects by plasma melting and spraying of the powdered alloys of the invention, has been proven feasible. The plasma-arc spraying technique further holds promise for preparing extremely rapid chilled, and thus very-fine-grained, alloying powders, which then can be consolidated into shapes by powder-metallurgical techniques.

It is importaNt in whatever manner of fabrication which is employed that the eutectic or near-eutectic liquid phase be rapidly cooled in order to assure the formation of the fine-grained, lamellar microstructure of the invention.

Dense bodies can also be prepared from powdered material by hot pressing, and also by cold pressing followed by sintering, preferably with the addition of sintering aids. The powders may comprise the desired carbides and metals. The iron group metals or their alloys, as well as manganese and cooper-containing alloys, may be used as sintering aids. Among these ,nickel or

nickel-iron alloys seem to afford the best properties in terms of toughness and shock-resistance, but as cutting tools, the sintered materials are inferior to the cast alloys.

ALLOYING OF THE GROUP IV*a* METAL (Ti, Zr, and Hf)—W—C BASE ALLOYS OF THE INVENTION

The ternary alloys from all three base systems of the invention can be extensively modified by alloying additions of other metals.

Alloying possibilities were determined by preparing cutting tools and using these on type 347 stainless steel. However, the observations are applicable to the composites in general.

1. Ti—W—C base tools had the best cutting performance in terms of tool life. The optimum composition in this base system lies at, or near, the composition Ti—W—C(23—52—25 atomic percent) which is slightly hypereutectic. Hypoeutectic alloys located to the tungsten side of the pseudobinary eutectic have slightly higher wear rates than the optimum composition, but also have somewhat higher edge-stability and cracking resistance, alloys located to the titanium side of the eutectic have good wear characteristics, but tend toward chip-welding at high cutting speeds; alloys with more than 28 atomic percent carbon are hypoeutectic, contain primary metal-phase, and are subject to high wear.

2. Tungsten may be partially replaced by molybdenum and or Rhenium (for instance, up to 20 atomic percent of the base alloy system) without impairing strength or abrasion resistance of the composite. Small quantities of chromium (up to 10 atomic percent of the base alloy system) also may be substituted for tungsten, but larger quantities result in embrittlement of the composites. In certain embodiments for certain useages all of the tungsten may be replaced by its alloying substituents. It must be remembered however, that in any substitution, the fundamental system must be maintained, namely the three elemental groups titanium and its alloying materials, tungsten and or its alloying substituents and carbon of the system. The same is true for the hafnium based systems and the zirconium based systems.

3. The Group IV*a* metals (Ti, Hf, and Zr) may be interchanged for each other in any ratio in their respective base alloy systems, for most purposes. Low level alloying (1 to 5 atomic percent) of the Ti—W—C system with Zr or Hf increases the tool life in comparison with unsubstituted base alloys, still higher concentrations of Hf or Zr result in a graidual drop-off of cutting performance to the levels observed for ternary Zr—W—C or Hf—W—C alloys. Generally speaking, the Hf or Zr will not be substituted in an amount in excess of 20 atomic percent of the Ti in the base alloy Ti—W—C system. Ti generally is not substituted in a Zr based system over 20 atomic percent for the Zr.

More typically, the alloying Group IVa metal or metals Hf and Zr will comprise not more than 5 atomic percent of the base alloy Ti—W—C system. However, in a Zr—Hf, and in a Hf—Zr based system a 50:50 ratio of these metals as the Group IV*a* metal component would not be uncommon. The 1 to 5 At. percent alloying substitutions of Zirconium or Hafnium, or mixtures of these two elements for Titanium in Ti—W—C base alloys significantly improve cutting tool wear resistance as well as abrasion resistance of these composites in other applications such as hard facings for plows, bulldozer blades, bearings and dies. It is believed that the small Zirconium or Hafnium additions reduce the friction between cutting tool and work piece leading to lower cutting temperature which results in greater tool life and improved overall performance; similarly, advantages in the economics of facings, bearings and dies are to be expected with the utilization of these Zirconium or Hafnium containing Ti—W—C composites by virtue of the improved abrasion resistance. In addition, these small Zirconium or Hafnium additions are believed to function as grain refiners of the Ti—W—C base alloy microstructure, presumably contributing to the overall strength of the composites.

4. Substitution of Group V*a* metals, such as vanadium for tungsten or titanium in quantities up to 10 atomic percent of the base alloy system may be made. In the composites this decreases the cracking sensitivity, but somewhat impairs abrasion resistnace. In a cutting tool the addition impairs performance and edge strength.

Up to 5 atomic percent of such Group V*a* metals as niobium and tantalum, each alone or in combination, can be added to replace the Group IV*a* metal. Edge chipping in a cutting tool is increased if the noibium-tantalum addition is greater than 5 atomic percent. Such additions up to and over 5 atomic percent may improve cratering and chip-welding characteristics in cutting tools. Overall, the addition of Group V*a* metals in quantites of more than 5 atomic percent (preferably, not more than 2 atomic percent) is not recommended for Vanadium, Niobium or Tantalum.

5. No significant change in cutting performance was observed upon substituting up to 10 atomic percent rhenium for tungsten. Substitution of rhenium up to 20 atomic percent for tungsten appears acceptable.

6. Low level additions of iron group metals (Co, Ni, Fe), of manganese and copper, and of rare earch metals in quantities less than 3 atomic percent of the carbide composite of the invention were found to be essentially inert, i.e., to have little or no effect on the physical and mechanical characterictis of the composites.

7. Eutectic, or slightly hypereutectic, Zr—W—C and Hf—W—C based alloys are tougher than Ti—W—C based alloys, but were found to have higher wear-rates in cutting tool applications.

The base alloy systems of the invention including the added amount of cutting tool performance improving alloying metals will typically comprise at least 90 atomic percent of the carbide composite. Generally speaking, the atomic percent age of inerts is held to less than 3 to 5 atomic percent of the carbide composite.

The refractory metal phase of the lamellar microstructure of the invention will typically have a melting point around 2,700°C which is a decided improvement over the 1,400°C melting temperature of the conventional cobalt cutting tool.

The rate of cooling of the alloy of the invention during its fabrication determines grain size. Desirably, cooling is accomplished at a rate of at least 20°C per second to obtain a generally fine grain. Cooling at a slower rate gives a product with a coarser grain. Preferably, cooling is performed at a rate of more than 30°per second.

Preliminary test results indicate transverse rupture strength levels for the cast Ti—W—C eutectic struc-

ture in the range of from 220,000 psi and extending to above 350,000 psi, depending upon fabrication conditions.

The majority of tests has been carried out in studying the performance of the alloys as cutting tools in straight turning of cylindrical test bars on a LeBlonde machineability lathe. For these tests, the carbide alloys were either machined into inserts suitable for clamping in standard tool holders, or more or less irregular shaped bits were brazed onto steel tool holders and then ground on a K.O. Lee diamond grinder to the desired geometry. The test material consisted of annealed 347 stainless steel in the form of 3 inch diameter × 18 inch long cylindrical bars. The surface was removed to a depth of 0.050 inch prior to testing the experimental alloys. In the standard test, the steel was cut at 400 surface feet per minute (sfm), using a depth of cut of 50 mils and a feed of 10 mils per revolution. The tool geometry for the standard test was as follows: back rake, 0°; side rake, 5°; side relief, 5°; end relief, 5°; side clearance end angle, 25°.

A number of representative commercial cutting were evaluated under machineability test conditions described above. In addition to the examples below, a selected list of additional tests is contained in Table II, Infra.

FIG. 3 graphically depicts the comparative wear curves obtained in the turning of Type 347 stainless steel with the cast alloy tools prepared according to the invention and with top grade C-2 and C-50 type commercial carbides. It will be seen that the cast alloyed tools of the invention have equivalent wear resistance to the top grade wear resistant C-50 tools. In addition, it has been shown that the tools of the invention have equivalent toughness to that of the C-2 tools. Therefore, the cast tools of the invention combine the best qualities of the tough C-2 tools and the wear resistant C-50 tools.

EXAMPLE I

A button of an alloy Ti—W—C (19-58-23 atomic percent) was prepared by arc melting in a nonconsumable electrode arc furnace under helium at ½ atmosphere pressure; the melt was allowed to solidify on the water-cooled copper hearth. Metallographic examination of the alloy showed very small amounts of primary monocarbide grains in an eutectic lamellar matrix. The average lamellae width of the eutectic structure was about one micron. The hardness was $R_A = 86$. The tool was brazed onto a mild steel tool holder, ground to the standard tool geometry,, and tested in turning 347 stainless steel with the standard conditions outlined above. The tool life, based on a flank wear of 0.016 inch was 45 minutes; the tool showed local wear (crater) of 0.028 inch at the end of the cutting flank.

EXAMPLE II

An alloy Ti—Zr—W—C (20.5-2.5-52-25 atomic percent) (standard alloy R1 in FIG. 3) was prepared in the same way as the sample described under Example I. The composite had a hardness of $R_A = 87$, and the metallographic examination showed small amounts of primary monocarbide in an eutectic matrix (substantially identical to the microstructure shown in photomicrograph of FIG. 2). The average lamellae width of the eutectic was about 0.4 microns. The heterogenous matrix of the photomicrograph of FIG. 2 is an eutectic of metal plus carbide, and the white or light islands are primary carbide. A uniform wera rate of 0.07 mils per minute was derived from a 40-minute turning test of 347 stainless steel with aforesaid standard conditions, yielding an extrpolated tool life of 190 minutes (0.016 inch flank werar). Cratering of the tool after 40 minutes cutting time was negligible.

EXAMPLE III

An arc cast alloy Hf—W—C (27-51-22 atomic percent) containing a small amount of primary carbide grains in addition to the eutectic lamellar microstructure was prepared. Tool life in the standard test on 347 stainless steel was 15 minutes, with the tool showing negligible cratering or edge wear at the end of the test.

EXAMPLE IV

The alloy cited under Example II and another arc cast alloy Ti—Hf—W—C (20.5-2.5-52-25 atomic percent) were tested for edge stability by gradually increasing the feeds while maintaining a surface speed of 400 feet/min. and a cutting depth of 0.050 inch. Both tools performed reliably at feeds up to 0.05. inch per revolution. At still higher feeds, the tool edges showed signs of chipping.

EXAMPLE V

The behavior of the cast carbide tooling materials at high depth of cut were established in another test run using the same alloys as listed under Example IV with a cutting speed of 400 sfm (surface feet per minute). A constant cutting depth of ¼ inch was maintained in the experiments, while the feed was gradually increased, starting at 0.005 inch per revolution. No breakdown occurred at feeds up to 0.030 inch/rev., after which the experiment had to be stopped for lack of lathe power.

EXAMPLE VI

An arc cast alloy Ti—W–C (19-58-23 atomic percent) was comminuted to a grain size below 50 microns and thoroughly mixed with 3 weight percent nickel powder. The mixture was cold-compacted at 4 tons/-square inch in steel dies and then sintered for 1 hour at 1,500°C under vacuum. The metallographic examination showed a dense structure consisting of rounded monocarbide grains embedded in a metallic matrix. Tool life in the standard turning test on 347 stainless steel was 14 minutes. The tool had a higher crater wear than the cast alloy of the same composition.

EXAMPLE VII

A composite tool was fabricated by facing one edge of an M-2 tool steel insert with a 0.080 inch wide × 0.20 inch long × 0.050 inch thick platelet of the cast standard alloy R1, Ti—Zr—W—C (20.5-2.5-52-25 atomic percent). The carbide tip was attached to the steel insert by brazing. The performance of this composite tool under the standard test condition on 347 stainless steel was found to be the same as the solid carbide inserts; however, as a result of the lower thermal conductivity of the tool steel base compared to the cast carbide alloys, higher tip temperatures, and, as a consequence, higher wear rates were observed on the composite insert as the total load on the tool was increased by either increasing the depth of cut or the feed.

TABLE II

Selected List of Test Data Obtained from Cast Ti(Zr,Hf)-W-C Alloy, and From Commercial C-2 and C-50 Grade Carbide Cutting Tools, in Turning Type 347 Stainless Steel

Alloy Composition, Atomic Percent	W_B	T_B	W_U	T_L	Remarks
Ti(26)-W(51)-C(23	5	3	0.14	65	slight welding tendency
Ti(21)-W(56)-C(23)	4	4	0.25	65	
Ti(20)-W(55)-C(25)	3	3	0.13	110	
Ti(23)-W(52)-C(25)	3	5	0.16	75	
Ti(26)-W(49)-C(25)	4	6	0.09	160	
Ti(25)-W(48)-C(27)	4	5	0.06	160	slight chipping tendency
Ti(22)-W(51)-C-27	4	4	0.10	110	
Ti(23.5)-Zr(2.5)-W(49)-C(25)	4	5	0.07	160	
Ti(17.5)-Zr(2.5)-W(55)-C(25)	4	4	0.16	75	
Ti(20.5)-Zr(2.5)-W52)-C(25)	4	8	0.07	190	alloy R1
Ti(20.5)-Hf(2.5)-W(52)-C(25)	5	8	0.08	130	alloy R2
Ti(23)-Ta(2.5)-W(51.5)-C(23)	7	5	0.19	65	slight chipping tendency
Hf(22)-W(60)-C(18)	6	5	0.44	25	
Hf(25)-W(55)-C(20)	4	4	0.25	50	
Hf(25)-W(48)-C(27)	4	4	0.25	50	
Zr(25)-W(55)-C(20)	7	4	0.70	20	
Zr(28)-W(47)-C(28)	6	5	0.30	40	
Commerical Alloys					
Carboloy 370 (C-50 Type)	3	3	0.08	160	
Carboloy 883 (C-2 Type)	—	—	1.10	10	

Legend:
W_B — Break-in wear, mils
T_B — Break-in time, minutes
W_U — Uniform wear rate, mils per minute
T_L — Extrapolated tool life in minutes for 0.016 inch wear

*Data shown are for specific heats of stainless steel. Results will vary from heat to heat for stainless steels of the same specification, comparative ranking will remain as shown.

What is claimed is:

1. A cast carbide quaternary composition of the elemental formula:

$$[(Ti_a)(Zr_b)]- W - [C_q]$$

which consists essentially of Titanium, Zirconium, Tungsten, and Carbon wherein

a is the *At.* percent of Ti

b is the *At.* % of Zr

q is the At.% of C

$$10 \leqq a + b \leqq 40$$

$$20 \leqq q \leqq 30$$

when the fraction $a/(a+b)$ is equal to or greater than ½

and

$$15 \leqq a+b \leqq 40$$

$$15 \leqq q \leqq 30$$

when the fraction $a/(a+b)$ is less than ½

the balance being W when $a/(a+b)$ is equal to, greater than, or less than ½.

2. A cast carbide quaternary composition of the elemental formula:

$$[(Hf_a)(Zr_b)]-W-[C_q]$$

which consists essentially of Hafnium, Zirconium, Tungsten, and Carbon wherein

a is the atomic percentage of Hf

b is the atomic percentage of Zr

q is the atomic percentage of C

wherein

$$15 \leqq a+b \leqq 40$$

$$15 \leqq q \leqq 30$$

the balance being tungsten.

3. A cast carbide composition of a higher order than quaternary of the elemental formula:

$$[(Ti_a)(Zr_b)]-[(W)(X)(Cr)(R)] -C_q$$

which consists essentially of at least Titanium, Zirconium, Tungsten and Carbon wherein

a is the atomic percentage of Titanium

b is the atomic percentage of Zirconium

q is the atomic percentage of Carbon

X is at least 1 element selected from the group consisting of Molybdenum and Rhenium, said X being present in the range of from 0 to about 20 At. percent;

Cr is present in the range of from 0 to about 10 At. percent

R is at least 1 element selected from the group consisting of Vanadium, Niobium, and Tantalum, said R being present in the range of from 0 to about 5 At. percent, wherein

$$10 \leqq a+b \leqq 40$$

$$20 \leqq q \leqq 30$$

when the fraction $a/(a+b)$ is equal to ½ or greater than ½ and

$$15 \leqq a+b \leqq 40$$

$$15 \leqq q \leqq 30$$

when the fraction $a/(a+b)$ is less than ½

the balance being w when $a/a+b$ is equal to, greater than, or less than ½; the balance being tungsten.

4. A cast carbide composition of a higher order than quarternary of the elemental formula:

$$[(Hf_a)(Zr_b)] - [(W)(X)(Cr)(R)] - C_q$$

which consists essentially of at least Hafnium, Zirconium, Tungsten and Carbon, wherein

a = At. percent of Hf

b = At. percent of Zr

q = At. percent of C

X is at least 1 element selected from the group consisting of Molybdenum and Rhenium, said X being present in the range of from 0 to about 20 At. percent;

Cr is present in the range of from 0 to about 10 At. percent;

R is at least 1 element selected from the group consisting of Vanadium, Niobium, and Tantalum, said R being present in the range of from 0 to about 5 At. percent and wherein

$$15 \leqq a+b \leqq 40 \text{ At.\%}$$

$$15 \leqq q \leqq 30 \text{ At.\%}$$

the balance being tungsten.

5. A cast carbide quaternary composition of the elemental formula

$$Ti—Zr—W—C$$

wherein

Ti is present at about 20.5 atomic percent

Zr is present at about 2.5 atomic percent

W is present at about 52 atomic percent

C is present at about 25 atomic percent.

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