Aug. 2, 1955 C. G. GOETZEL 2,714,245

SINTERED TITANIUM CARBIDE ALLOY TURBINE BLADE

Filed Dec. 7, 1951 3 Sheets-Sheet 1

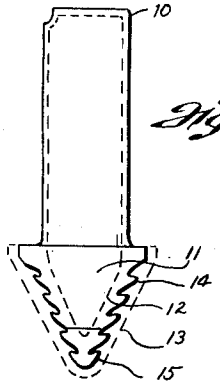


Fig. 1.

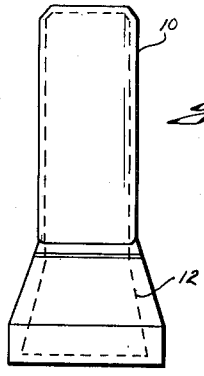


Fig. 2.

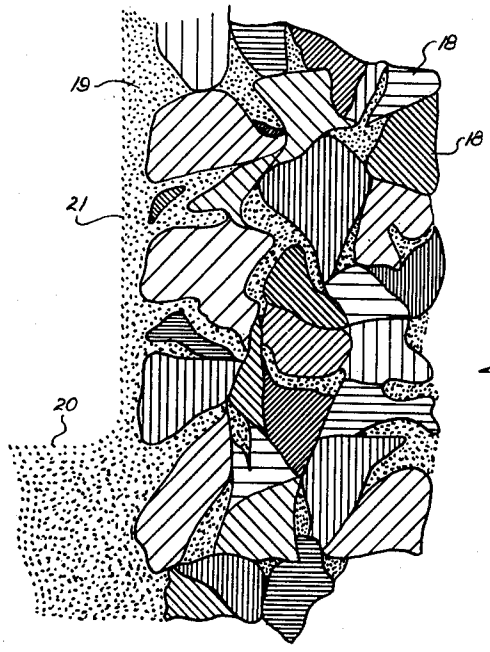


Fig. 3.

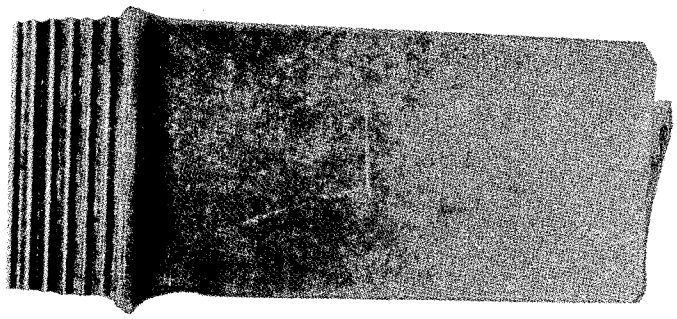
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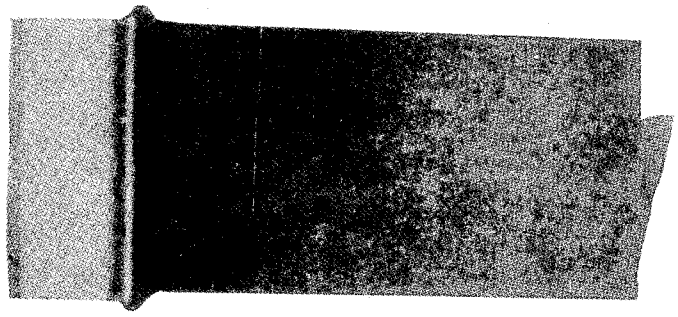
SINTERED TITANIUM CARBIDE ALLOY TURBINE BLADE

Filed Dec. 7, 1951 3 Sheets-Sheet 2

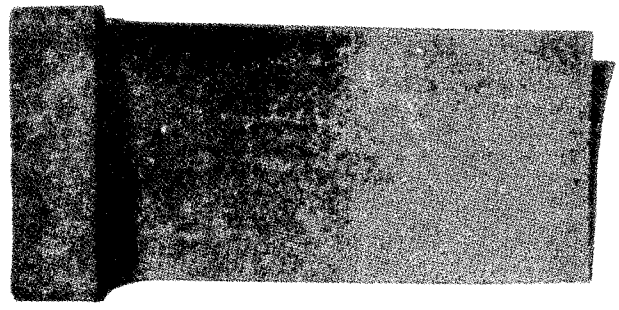
Jet Engine Turbine Bucket from Infiltrated Titanium Carbide



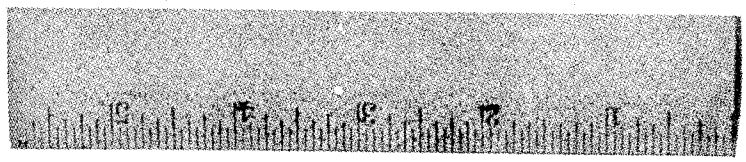
Bucket with Machined Root
FIG. 6



Infiltrated Bucket
FIG. 5



Sintered Preform
FIG. 4



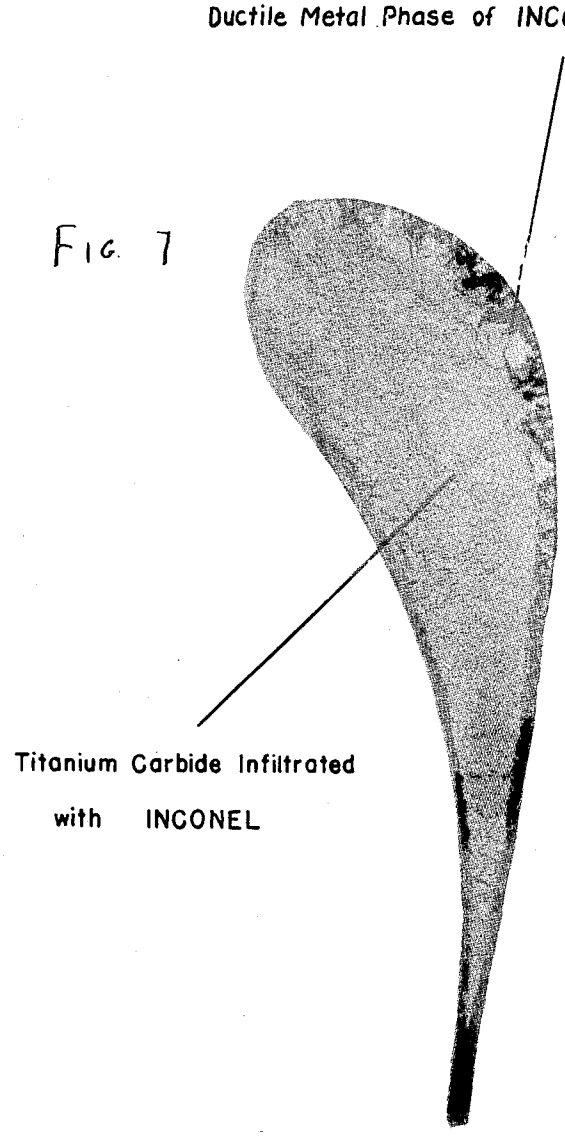
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SINTERED TITANIUM CARBIDE ALLOY TURBINE BLADE

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FIG. 7



Jet Engine Turbine Bucket
Cross Sectional Fracture
3 Diameters magnification

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United States Patent Office

2,714,245

Patented Aug. 2, 1955

2,714,245

SINTERED TITANIUM CARBIDE ALLOY TURBINE BLADE

Claus G. Goetzel, Yonkers, N. Y., assignor to Sintercast Corporation of America, Yonkers, N. Y., a corporation of New York

Application December 7, 1951, Serial No. 260,504

5 Claims. (Cl. 29—182.8)

This invention relates to composite material shaped turbine blades for use at an elevated temperature.

Sintered refractory metal carbides and particularly titanium carbide bonded by certain of the transition group VIII elements have for many reasons become of great interest to the jet engine producers, and particularly to producers of turbine blades.

Carbides are generally strong at high temperatures and do not lose strength with increasing temperatures as readily as metals and alloys do. The use of cobalt, nickel or iron base alloys nowadays generally used for jet engine blades is limited to a temperature of about 1500° F.–1650° F. for high stress application. The oxidation and corrosion resistance of these alloys, particularly when containing higher percentages of chromium, provides a limited protection up to about 1800 to 2200° F.; however, the load carrying capacity is restricted to about 1500 to 1600° F.

Titanium carbide should be preferred for the making of the turbine blades on account of its light weight, since in turbine blades, where centrifugal stresses are very high due to the high rotational speed of the turbine rotor, the stresses decrease with the decrease of the specific gravity of the blade material. Titanium carbide is also low in strategic alloy content.

However, titanium carbide has an unusual susceptibility to damage by thermal shock. While cemented titanium carbide shows better thermal shock resistance than almost all other intermetallics (intermetallic compounds cemented with a metal binder), it is still inferior to the regular metallic alloys.

In addition, all types of intermetallics are seriously damaged by small surface imperfections, such as cracks, which act as stress-raisers, and by grain boundary oxidation, which also results in stress concentration effects. Under these conditions titanium carbide cannot withstand repeated temperature or stress cycles, nor can it withstand bending type strains.

In view of the above and in spite of its advantages, titanium carbide has not as yet been successfully applied to the production of turbine blades for jet engines, although many experts have labored for years with this specific problem.

Among the difficulties causing the above mentioned blade failure are:

Thermal or mechanical shock.—This is particularly grave since the failure of a single blade will destroy all

the remaining blades on account of the low ductility of titanium carbide.

Fatigue failures.—These failures occur because of surface defects, such as notches, tiny surface cracks, intercrystalline oxidation. The failures usually originate at the fine trailing edge of the blade which must withstand the greatest deformation resulting from bending stresses imposed on the blade. Once a crack or notch exists, failure by fatigue proceeds rapidly.

Oxidation or corrosion.—Corrosion or oxidation of the grain boundaries results in a notch or stress concentration effect leading to early fatigue failures, cracking, fracturing and general breakdown of the blade attachment and turbine disc weakening.

Probably the main objection to the use of sintered titanium carbide turbine blades is that no reliable and durable attachment means are available for mounting the composite material blades, and particularly buckets and nozzle vanes, on jet turbines. The nozzle vanes are usually attached to an inner and outer shroud ring which, in turn, is fastened to a stationary part of the turbine, whereas the buckets are connected to the rotating turbine wheel. No suitable attachment means are so far available for either type of turbine blade elements from titanium carbide. Moreover, it is common practice when attaching turbine blades to provide a definite but small amount of looseness of fit to improve the damping capacity of the blade against destructive vibrational effects.

I shall enumerate some of the difficulties encountered in the employment of titanium carbide for turbine blades, buckets or vanes for jet engines.

Inability to fasten the blade without cracking of the attachment and failure of the blade.

Thermal shock failure, which results in cracks.

Fatigue failures, caused by thermal shock cracks, and also by cracks in the brittle trailing edge produced, in turn, by surface defects, outside penetrations, nicks.

Loosening of the blade due to creep caused by heat developed at the attachment points of sintered carbide blades.

The object of the present invention is to produce titanium carbide turbine blades infiltrated with a ductile and high temperature corrosion-resistant cobalt, nickel or iron base alloy. The use of an excess of the ductile infiltrant metallic alloy acts to counterbalance the above listed difficulties, as described in the above-mentioned United States patent application Ser. No. 795,102.

The presence of a chance excess of the metallic infiltrant phase will, however, not solve the various problems, which require specific controls.

First, there is the problem of a deliberate, accurate and reproducible provision of an excess of ductile metallic alloy. Second, there is the problem of equally deliberate provision of excess metallic alloy in specific locations in the blade. This calls for scientific prorating of carbide and of ductile metal alloy to provide for protection against high stresses in one instance, for thermal shock in a second instance, and for blade fastening in a third.

In the root of the blade, for example, considerable excess ductile metal alloy phase must be provided for machining the fir tree root with enough left over to pro-

vide a ductile metallic seat in the root attachment to avoid the previously noted cracks, crumbling and general breakdown. The use of excess alloy at the root will decrease the quantity of heat conducted to the wheel rim and will permit cooler rims, less creep, safer blade attachment.

The carbide and metal phase at the trailing edge must be carefully proportioned so as to guard against fatigue type bending stresses. The carbide skeleton phase must not be contained in those extreme portions of this thin trailing edge where its thickness is less than about 1/32 inch (see Fig. 7). Only the ductile metal alloy infiltrant in this location can readily withstand thermal shock and fatigue stresses, whereas the carbide containing composite material would tend to fail in this location due to its lower ductility, shock resistance, and fatigue resistance. The bulk of the blade proper, however, must be the carbide phase to provide load carrying capacity.

On the curved blade faces, on the other hand, only a slight film of ductile metal alloy phase will be necessary to permit more perfect surfaces which are not susceptible to stress concentration factors, such as cracks, nicks, etc. Possibly a film of molecular thickness would answer the purpose, but in practice the film should be adequate to cover the surface irregularities of the air foil section, say one-half to ten thousandts of one inch thick.

Moreover, the thickened surface coating of the infiltrant ductile metal alloy, serves firmly and reliably to fasten the blade, and also to permit machining and heat transfer blocking, as well as offering the desired oxidation resistance.

The practicability of fastening is enhanced by the ductility of the infiltrant alloy with which the end facings are enriched. This will prevent misalignments and maladjustments of the blades. Also, overheating due to the high heat conductivity of the metal carbide skeleton bodies is eleminated because the heavy infiltrant alloy-enriched end sections of the blade have relatively low conductivity and, therefore, act as a barrier to heat flow into the shroud rings in the case of the vanes, or the wheel in the case of the buckets. Fastening is facilitated by the circumstance that the coefficient of expansion of the end section enriched by the infiltrant alloy matches more closely that of the alloy shroud rings or turbine wheel than would the carbide-base body of the blade.

Since titanium carbide is light, its use in turbine components of aircraft power plants will effect important weight savings which go beyond the actual weight reduction in the turbine elements. This is so, because structures of engine mountings, frames, etc., may be lightened, thus increasing the load carrying capacity of the aircraft. Moreover, in the buckets, where centrifugal stresses are very high due to the rotational speed of the turbine rotor, the stresses are reduced with the specific weight of the blade material whereby the safety requirements with regard to the attachment of the blades are more successfully met if titanium carbide is used for the production of the blade skeletons.

A superior and complete solution of the problems concerning the attachment of turbine blades of sintered and infiltrated titanium carbide is, according to the present invention, accomplished by a combination of thickening the attachment portions of the blades, with the use of a particular type of titanium carbide for the production of the blade skeletons.

The increase in the thickness of the infiltrant alloy layer may vary depending on the blade type; the layer must be much thicker, possibly as high as one inch, on buckets, where deep serrations have to be produced to connect with the rotating turbine wheel. Two thickened infiltrant sections must be provided on each end facing of nozzle vane blades. Since the ends of the nozzle vanes are not machined for the purpose of producing deep serrations as in the fir tree end of the buckets, but serve only to

provide a secure fit into the slots of the shroud rings, the infiltrant enriched end layers need not be more than about one-fourth of an inch thick at each end facing.

As already mentioned before, the important feature of this invention consists in the combination with the thickening of the infiltrant metal layer at the attachment places of the turbine blades, buckets and nozzle vanes, of a particular type of titanium carbide for the production of the blade skeletons.

Articles having TiC as skeleton material and major volumetric constituent may be used at temperatures of 1650 to 2000° F. and above and will, therefore, be suitable for the manufacture of turbine blades. Furthermore, titanium carbide has a density of only 4.9 grams per cc., or about 60% of that of the conventional type of superalloys used for turbine bladings, buckets, nozzle vanes in jet engines. This is obviously most important for components subjected to centrifugal force and allows diminution of the cross-section of engine parts subjected to high temperature.

Titanium carbide is a very high melting, very hard compound which compares in its physical properties, including its strength at normal and elevated temperatures, to other refractory carbides, e. g. tungsten carbide and tantalum carbide. Since the absolute values of its strength are about the same, but the density of TiC is only about 1/3 of tungsten carbide and 1/2 of tantalum carbide, the strength-weight ratio is much more favorable than in either one of the other materials.

It has been well known in the art that the method of forming composite structures of a high melting substance and a low melting alloy phase by infiltrating the latter into a sponge-like skeleton shape of the high melting component produces articles of a high degree of density, uniformity and soundness, without undergoing such drastic changes in shape and dimensions, e. g. shrinkage, as are encountered when a mixture of the high melting and low melting constituents is compressed into the shape of a blade and then subjected to conventional sintering techniques. It is also a well-known fact that the infiltration process produces composite structures, whose high melting skeleton phase and lower melting infiltrant metal phase form two closely interwoven or intertwined networks having unusual physical properties derived from cumulative predominant properties of the two phases. The infiltration technique permits ready production of composite bodies having an ideal combination of chemical and physical properties, in which the volumetric ratio of the two phases may vary over a much wider or a different range, e. g. 50/50 to 70/30 volume ratio of titanium carbide and metal alloy, than would be possible or practical by the conventional powder metallurgy methods of cold pressing and sintering, whose concurrent shrinkage effects cause limitations of the range, or hot pressing, whose concurrent loss of binder by sweating out also causes limitations of the range.

For the foregoing reasons, the application of infiltration technique to the production of composite turbine blades having TiC as the major or skeleton forming constituent is obviously desirable.

However, infiltration of skeleton bodies produced from TiC powder with lower melting metals or alloys cannot readily be accomplished. Pure TiC has little chemical affinity or wetting property towards molten iron, nickel and cobalt, and its solubility in these metals at elevated temperatures, i. e. up to 1600° C., is also limited. Therefore, infiltration with the aid of capillary action is difficult to accomplish, and not even very small cross-sections can be penetrated by the liquid infiltrant, except in regions strictly limited to the surface in contact with the infiltrant.

It is, therefore, an important object of the invention to remedy such deficiencies of the TiC and to provide operative conditions which permit a complete penetra-

tion of the pores of the TiC blade skeletons for the purpose of a reliable support of the infiltration layers.

As later explained more in detail, the presence of additional elements in the TiC powder is of considerable importance. One of these elements forms a gaseous and volatile reaction product which, upon escape from the forming skeleton, leaves channels and junctures between the pores. The other addition, namely the element cobalt, nickel, or iron in amounts from 1–10%, forms a strong and metallic cementing binder element which imparts in the titanium carbide skeleton body sufficient coherence to withstand the eruptive forces during infiltration with the molten heat resistance alloy without excessive swelling, deformation, warpage, twisting or cracking of the skeleton body during infiltration or during solidifying of the infiltrated body.

A subsequent heat treatment under carefully controlled atmospheric conditions further aids the strengthening of the TiC skeleton body. In the case of a cold-pressed, rampacked and dynamically loaded skeleton, this treatment constitutes a sintering operation of the coated TiC particles, the coating performing the function of a binder during pressing and of a cement during sintering. By carefully timing the heat treatment and employing a temperature low enough, for instance 1400–1550° C. to permit only superficial solubility between the TiC and the metal or alloy phase and reducing the treatment to a period of about ½ to 1½ hours, shrinkage of the skeleton bodies can be kept in the order of zero to two per cent and the interconnected pore system can be preserved. In this case, the presence of a pore-channel creating element, such as carbon, is of importance. In spite of this limited consolidation and shrinkage, the cohesive strength of the skeleton is augmented manyfold by this treatment.

Similar heat treatments, for instance, sintering, performed on hot-pressed TiC skeletons also has a strengthening effect. Furthermore, the treatment serves to clean the skeleton body of superficial impurities originating in the hot-press operation, e. g. adherent carbon, mold lubricant, etc., thereby freeing the pore channels at the surfaces. It also serves for the removal by volatilization of any remainder of additional pore channel creating element. The heat treatment of hot-pressed TiC skeletons is preferably carried out at somewhat higher temperatures, e. g. 1450–1650° C.

In the sintering heat treatment, the cold-pressed or hot-pressed TiC skeletons are advantageously submerged inside a suitable vessel in an inert powder pack, such as a stable refractory oxide, for instance, alumina, magnesia or beryllia, or a degasified type of carbon, for instance, Norblack, Thermatomic Carbon. Highly desiccated hydrogen furnace atmosphere is passed through the pack and permitted to contact and penetrate the TiC skeleton bodies. Other atmospheres found suitable are helium or argon at atmospheric pressure of the same gases as well as hydrogen or nitrogen in a partial vacuum at a subatmospheric pressure of 50–250 microns mercury column.

The latter procedure is preferred where alloy infiltrants are used which have a strong tendency to react with, or absorb or dissolve gases, especially those given off by the titanium carbide skeleton during infiltration. Certain nickel-chromium alloys fall in this lass. In order to achieve a maximum elimination of gases in the titanium carbide, the sintering treatment in partial vacuum is preferably conducted at 1500–1650° C. in the case of cold pressed titanium carbide skeletons, and at 1600–1800° C. in the case of hot pressed titanium carbide skeletons. This treatment increases the shrinkage to two to five percent, and augments the cohesive strength of the skeleton further, mainly because of a partial volatilization and interstitial re-deposition of the metallic addition element previously added to the titanium carbide.

The selection of the type and grade of the TiC powder is, as mentioned before, extremely important for the production of skeletons that will be strong and coherent during heat treatment and adapted properly to support infiltrant alloy layers of considerable thickness.

It was found that a TiC powder, rich in free carbon, and containing for instance 20.0–21.5 total carbon and about 1.0–3.0 free carbon satisfactorily serves the purposes of this invention as it enables the production of a skeleton body, which is a reliable support for a thickened infiltrant alloy layer; this is particularly so if the powder is subjected to a refining and purification heat treatment in pure hydrogen or other non-oxidizing atmosphere at 1900–2300° C. whereby skeleton bodies can be produced having such strength and coherence as to eliminate deformations, volumetric changes or local stress concentrations causing cleavages and cracks. Such skeleton bodies will not undergo swelling and gross volume increases upon infiltration, even though the amount of infiltrant may be uncontrolled and considerably in excess of the value calculated to fill all pores of the skeleton.

While I believe that the free carbon content of the TiC powder is the critical factor, it is possible that some other substances have contributed to the good results achieved. The following is the analysis of TiC powder which I found unsuitable for the object of this invention.

	Per cent
Ti	77.1
Total C	17.36
Free C	0.07
N	0.85
Si	0.21
Fe	0.69
Al	0.15
Ca	0.10
V	0.60
Zr	0.70
O	Approx. 2.00

Lesser impurities are B, Mn, Mg, Cb, Cu, Ba and Na.

The following is the analysis of TiC powder which I found suitable:

Ti	[1] 75.2
Total C	[1] 21.0
Free C	[1] 3.17
N	[1] 0.70
B	[2]
P	[2]
Si	0.2
Fe	0.6
Al	0.07
Mg	0.002
Ca	0.005
V	0.3
Zr	0.8
Cb	0.03
Hf	[2]
Cr	0.02
Mn	0.01
Ni	[2]
Cu	0.005
Sn	0.001
Zn	[2]
Sb	[2]
Pb	[2]

[1] By chemical analysis; others by spectrographic analysis.
[2] Not detected spectrographically.

The physical properties of test bars made of titanium

carbide powder of the above designated type are given in the following Table I.

TABLE I

Physical properties of infiltrated titanium carbide composites from high free carbon-containing titanium carbide powder

	Transverse Rupture at Room Temp., p. s. i.	Transverse Rupture at 1,800° F., p. s. i.	Deflection under Break Load at 1,800° F., in.
Skeleton Composition: TiC-Ni Binder, 90–10 weight ratio:			
Low TiC Concentration (approx. 60% by Vol.) Nichrome infiltrant	83, 700–216, 000	60, 400–82, 700	0. 060–0. 238
High TiC Concentration (approx. 75% by Vol.) Nichrome infiltrant	79, 500–138, 000	78, 400–88, 100	0. 035–0. 228
Skeleton Composition: TiC-Co binder, 90–10 weight ratio:			
Low TiC Concentration (approx. 60% by Vol.) Stellite #21 Infiltrant	74, 200–188, 500	115, 000–167, 700	0. 058–0. 215
High TiC Concentration (approx. 75% by Vol.) Stellite #21 Infiltrant	83, 000–141, 900	91, 000–149, 300	0. 031–0. 200

That these high qualities are mainly attributable to the presence of the free carbon is indicated by experiments of the inventor converting the high free carbon grade of the TiC into a low free carbon grade by alternate hydrogen and vacuum treatments at 2300° C. The resultant structure had much inferior physical properties than specimens made from the high free carbon containing powder.

The beneficial effects caused by the presence of from about 1 to about 3% free carbon in the TiC powder will now be summarized.

The carbon originating from the carburization of Ti dioxide by solid or gaseous reagents is present in a finely divided form, and uniformly dispersed over the surfaces of the TiC particles. The carbon acts, therefore, as an ideal lubricant and facilitates both cold and hot pressing. It makes possible the pressing of strong, coherent skeleton bodies of much lower density, i. e. volume concentration of TiC, than would be possible with the low free carbon containing powder, the lowest practical density being about 55% as against 65–70% for the other type of powder. This comparatively low skeleton volume concentration of TiC results in an optimum combination of resistance to oxygen, resistance to mechanical and thermal shocks, and strength and ductility at elevated temperatures.

In hot-pressing, part of the free carbon combines with the gases entrapped in the TiC powder, e. g. oxygen from the air or hydrogen from the controlled furnace atmosphere, to form carbon monoxide, hydrocarbons, or other carbon containing gaseous products. These gases tend to escape from the compact as the temperature of hot-pressing rises, thereby leaving behind channels which connect the different pores; this, in turn, facilitates rapid and complete infiltration allowing a minimum of time for the effects of a harmful solubility, stress concentration, and local distortion. The production of an interconnected pore structure with the aid of the carbon containing gaseous reaction products cannot be completed during the short hot-pressing, i. e. 1 to 10 minutes, but will be completed during the subsequent heat treatment, i. e., sintering, of the hot-pressed skeleton. An analogous process takes place in cold-pressed skeletons that are subsequently sintered.

When comparing the strength and the behavior of the skeletons of the low free carbon and high free carbon containing titanium carbide powder during infiltration, it becomes obvious that the increased strength produced by the cementing of the cobalt, nickel or iron binder during sintering is originally overshadowed most markedly by a bonding or alloying process caused by the presence of free carbon in the TiC powder. This is especially noticeable when cobalt is the binder metal and cobalt base Stellite and Vitallium alloys are the infiltrant. From the constitution of the system cobalt-carbon, it may be concluded that a considerable amount of carbon is dissolved by cobalt at elevated temperatures; this may amount to as much as 1.1% at the eutectic temperatures of 1300° C. In addition to the primary Co-solid solution, in which 0.1% is dissolved, and free carbon, the Co—C eutectic is present in the structure. Since it is known that in the molten state molecules of the compound Co_3C with 6.35% C are stable, and this cobalt carbide compound decomposes only slowly upon slow cooling and solidification, it must be assumed that the markedly increased rigidity of the cobalt-cemented TiC-base skeleton at the instance and temperature of infiltration is caused by the presence of the cobalt-carbide. Microscopic examination of skeleton bodies and infiltrated composites of the high free carbon-type powder has shown foreign phase border zones on TiC particles. This foreign phase was found to be hard and tenacious and of similar character as the TiC particles themselves, and it evidently constituted the binder metal with part of the free carbon contained therein, as cobalt carbide.

In the Ni—C system, the stable nickel carbide compound Ni_3C with 6.38% C exists in the molten state and down to about 1600° C., i. e. the upper temperature limit used during sintering; at lower temperatures down to 1400° C., the nickel carbide compound decomposes to a considerable degree. Therefore, the strength of the skeleton body during infiltration cannot be influenced by the presence of carbon in the nickel binder, to the same extent as in the case of the cobalt binder.

In the Fe—C system, the stable iron carbide compound Fe_3C is readily formed in the solid state by diffusion of free carbon into iron above 1300° C. The strength of the iron-cemented titanium carbide body during infiltration is definitely increased by the presence of carbon in the iron binder analogous to the case of cobalt-cemented titanium carbide skeleton.

According to experimental evidence, optimum physical properties are obtained when a high free carbon containing powder is subjected to a refining heat treatment in dry hydrogen atmosphere at 1900–2300° C. Mere heating of the powder in a sealed carbon container, without the use of any additional agents or elements, results in a reduction of the free carbon content by about ¼–½, e. g. from 2.2 to 3.3 to 1.5 to 1.0% and a corresponding increase of the combined carbon content, resulting in practically no change of the total carbon content of 20.0 to 21.5.

The following is an analysis of a satisfactory TiC powder before and after heat treatment:

	Before heat treatment	After heat treatment
Ti	[1] 76.00	[1] 77.20
Total C	[1] 20.20	[1] 20.09
Free C	[1] 2.26	[1] 1.46
B	0.0004	0.0004
Si	0.16	0.13
P	[2]	[2]
Zr	0.1	0.1
Sb	[2]	[2]
Fe	[1] 1.54	[1] 0.98
Cr	0.005	0.02
Mn	0.008	0.001
Pb	[2]	[2]
Mg	0.008	0.004
Sn	[2]	[2]
Cb	0.05	0.05
Al	0.07	0.07
Ca	0.03	0.03
V	0.3	0.3
Cu	0.001	0.0002
Zn	[2]	[2]
Ba	0.005	0.0005
Sr	[2]	[2]
Na	0.005	0.005
Li	0.0005	[2]
K	0.01	[2]
Ni	[2]	[2]
Mo	[2]	[2]
Co	[2]	[2]
Ce	[2]	[2]
N	[1] 0.59	[1] 0.73

[1] By chemical analysis; others by spectrographic analysis.
[2] Not detected spectrographically.

Moreover, such treatment burns out any residual impurities and will decompose and reduce any TiO_2 present in the original powder and convert the resulting Ti metal into TiC. Accordingly, the amount of free Ti is reduced by ¼–¾ and thereby any adverse, e. g. embrittling, effects connected with the alloying tendency of the metallic Ti with the metal binder during sintering and infiltration are also considerably reduced, or entirely eliminated.

These reduction and carburization reactions may be the reason why the physical properties of the heat treated and refined high free carbon containing TiC powder base materials are about 15–30% superior in hot strength and hot ductility to material made from the non-heat treated high free carbon grade powder. Another reason may be that an excessive amount of free carbon, e. g. about 2.0% or more may result in agglomeration of the carbon in the skeleton, which may in turn prevent the complete and uniform penetration of the infiltrant and produce gas pockets or other regions of mechanical weakness. A comparison of the physical properties obtained with cobalt alloy infiltrated TiC made from non-heat treated and heat treated high carbon containing TiC powder is given in the following Table II:

TABLE II

Physical properties of infiltrated titanium carbide composites from high free carbon-containing titanium carbide powder [1]

NOT-HEAT TREATED VS. HEAT TREATED TITANIUM CARBIDE POWDER

	Transverse Rupture at Room Temperature	Transverse Rupture at 1,800° F., p. s. i.	Deflection Under Break Load at 1,800° F., in.
Non-Heat Treated Titanium Carbide	92,300	108,500	.072
	119,900	108,500	.106
	164,900	117,200	.147
	125,900	126,000	.121
Average	125,750	115,050	.112
Heat Treated Titanium Carbide	110,400	124,000	.112
	142,000	127,000	.130
	172,200	132,000	.218
	166,000	130,500	.197
Average	145,150	128,400	.164

[1] Skeleton Composition, TiC-Co binder, 90–10 weight ratio. Low TiC concentration (approx. 60% by volume), infiltrated with Stellite #21 (Vitallium).

As infiltrant alloys, the following compositions have been found to be of particular usability.

Cobalt base alloys:
- "Haynes Stellite" Alloy #21 (Vitallium) (about 62% Co, 28% Cr, 5.5% Mo, 2.5% Ni, 1% Fe, 0.3% C max.);
- "Haynes Stellite" Alloy #31 (about 55% Co, 25% Cr, 10% Ni, 7% W, 0.6% Fe, 0.5% C max.).

Nickel base alloys:
- Nichrome–V (about 80% Ni, 20% Cr);
- Nimonic–80 (about 75 Ni, 21% Cr, 2.5% Ti, 0.7% Fe, 0.1% C max.);
- Inconel (about 78% Ni, 14% Cr, 7% Fe, 0.2% Cu max., 0.1% C max.);
- Nichrome (about 58% N, 22% Fe, 16% Cr, 3% Mn max., 1% Si, 0.25% C max.); and
- "Hastelloy" Alloy C (about 56% Ni, 17% Cr, 17% Mo, 5% W, 5% Fe, 0.15% C max.).

Iron base alloys:
- Stainless Steel Alloy #302 (about 72% Fe, 19% Cr, 9% Ni);
- Stainless Steel Alloy #310 (about 52% Fe, 25% Cr, 20% Ni, 1.5% Si, 1.5% Mn, 0.1% C max.);
- Stainless Steel Alloy #430 (about 82% Fe, 18% Cr, 0.1% C max.);
- Alloy "Multimet" (about 33% Fe, 22% Cr, 20% Ni, 19% Co, 3% Mo, 3% W, 1% Cb plus Ta, 0.2% C max.).

The invention will now be described with reference to the attached drawings. In these drawings:

Fig. 1 is a side view of one form of a turbine bucket produced in conformity with the invention, the bucket skeleton being made from a high carbon containing titanium carbide powder;

Fig. 2 is a front view of the bucket;

Fig. 3 is an enlarged showing of a section of the body at a place where the blade is attached to the turbine wheel;

Fig. 4 is a side view of the sintered skeleton preform;

Fig. 5 is a side view of the infiltrated bucket with its infiltrant alloy-enriched root portion;

Fig. 6 is a side view of the finished bucket with machined fir tree root; and

Fig. 7 is a cross-sectional view of the bucket clearly indicating the infiltrant alloy shell.

The blade has tip portion 10 and a base portion 11, which is machined to fit accurately into a suitably shaped notch or aperture in the turbine wheel.

It is desirable that the materials and dimensions of the base portion relative to the turbine wheel be chosen so that the blade will remain tightly held under the elevated temperature conditions of operation.

In the formation of the blade, the skeleton or high refractory phase dimensions or outline are diagrammatically indicated by the inner dashed lines 12 (Figs. 1, 2). The mold used during impregnation is arranged so that the infiltrant metal or corrosion resistant phase will fill the mold and form the outside of the tip portion of the blade and will reach the dashed lines 13 of the base 11. Thus there will be an excess of the infiltrant metal at 14 which can be machined to the desired base shape 15. Such a formation of the composite article with a thickened infiltrant metal portion may be termed "undercasting." As previously mentioned, the layer of infiltrant metallic material can be increased as desired at other points such as where erosion takes place on the blade.

In Fig. 3, the skelton or high refractory phase is indicated at 18, the particles thereof being formed together with intercommunicating pores therebetween, said phase being continuous.

The following examples illustrate the application of the invention in practice.

EXAMPLE 1

A free carbon containing titanium carbide of a 325

mesh size and containing 75.96 per cent Ti, 18.00 per cent combined C and 2.53 per cent free C was charged into a graphite crucible and heat treated in a dry hydrogen atmosphere at a temperature of 2300° C. for a period of thirty minutes. An agglomeration took place of the powder; the agglomerated mass was crushed and passed through a 140 mesh screen.

10 grams, constituting 10 weight per cent of cobalt powder of a 325 mesh size, was mixed with 90 grams of the titanium carbide powder; the mixture was fine-disintegrated in a stainless steel ball mill for twenty-four hours.

117.5 grams of the titanium carbide-cobalt powder mixture was charged into a graphite mold having a cavity corresponding to the curved air foil part of a jet engine turbine bucket, and providing an extension for a core section protruding into the root part of the bucket. The mold cavity was about 1/64 inch undersize all around, as compared with the dimensions of the ultimate product. The powder mixture was hot-pressed in the mold under a pressure of 1500 p. s. i. at a temperature of 1650° C. into a bucket-shaped skeleton body having a density of about 60 per cent of full. After cooling, the bucket was sintered in a graphite boat for about two hours at a temperature of 1700–1800° C. in a dry hydrogen atmosphere of a sub-atmospheric mercury pressure of 50 to 150 microns.

The skeleton body was now put into a form made of Alundum cement and located in a graphite carrier, the form having a cavity corresponding to the convex contours of the ultimate bucket product; the pocket of the cavity provided for the root was left with smooth walls and bottom. Its walls were sufficiently extended upward to provide a cavity for undercasting and infiltrating the skeleton core section of the root portion, and furthermore, for receiving the necessary allotment of alloy shot.

215.0 grams of Stellite #21 infiltrant alloy was then contacted with the skeleton body, in a manner that 103.0 grams of 1/16 inch thick sheet of the alloy was placed in contact with both curved faces of the air foil section of the bucket, while 112.0 grams of 1/16 inch granular shot of the alloy was placed in contact with the extension of the bucket reaching inside the smooth-walled pocket of the cavity.

Infiltration of the titanium carbide-base skeleton body and simultaneous filling of the empty space between the undersize skeleton bucket and extension and the mold cavity and pocket with the Stellite #21 alloy was then accomplished by heating in a high frequency heated carbon tube furnace under dry hydrogen atmosphere to 1500–1530° C., evacuating to a partial pressure of 100–200 microns while maintaining the temperature for two hours, and cooling under a pressure of 100–200 microns of dried nitrogen.

The density of the infiltrated, infiltrant-enveloped product was 6.93 g./cc., and the total weight 332.5 grams. The infiltrated product was finally subjected to a machining operation producing the fir tree type serrations in the surface of the infiltrant alloy-filled root part of the bucket.

EXAMPLE 2

The procedure as described in Example 1 was changed in so far as the infiltrant used was the alloy Stellite #31. The weight of the skeleton from titanium carbide-cobalt powder was 119.7 grams, that of the infiltrant alloy 220.2 grams. The temperature of infiltration was 1470–1500° C. The final density of the infiltrated, infiltrant-enveloped bucket was 7.05 g./cc., and the total weight 339.9 grams,

EXAMPLE 3

The procedure as described in Example 1 was changed in so far as the metal mixed in powdered form with the titanium carbide was carbonyl nickel, and the infiltrant used was the alloy Hastelloy-C. The weight of the skeleton from titanium carbide-nickel was 122.0 grams, that of the infiltrant alloy 226.0 grams. The temperature of infiltration was 1380–1410° C. The final density of the infiltrated, infiltrant-enveloped bucket was 7.23 g./cc., and the total weight 348.0 grams.

EXAMPLE 4

The procedure as described in Example 3 was changed in so far as the infiltrant used was the alloy Nichrome-V. The weight of the skeleton from titanium carbide-nickel powder was 119.0 grams, that of the infiltrant alloy 218.7 grams. The temperature of infiltration was 1460–1490° C. The final density of the infiltrated, infiltrant-enveloped bucket was 7.03 g./cc., and the total weight 337.7 grams.

EXAMPLE 5

The procedure as described in Example 3 was changed in so far as the infiltrant used was the alloy Inconel. The weight of the skeleton from titanium carbide-nickel powder was 119.7 grams, that of the infiltrant-alloy 220.1 grams. The temperature of infiltration was 1450–1480° C. The final density of the infiltrated, infiltrant-enveloped bucket was 7.05 g./cc., and the total weight 339.8 grams.

EXAMPLE 6

The procedure as described in Example 1 was changed in so far as the metal mixed in powdered form with the titanium carbide was carbonyl nickel, and to the mixture was added, as binder 2% of Resinox, a phenolic plastic product, dissolved in a small volume of acetone to disperse the binder and moisten the powder. The powder was then pressed at 20 p. s. i. in a steel die having the contours of the final bucket except for approximately 1/64 inch all around undersize dimensions and a smooth-walled undersize cavity for the root section. The compacted skeleton body was ejected and sintered in a graphite boat at a temperature of 1700–1800° C. for about two hours, in a dry hydrogen atmosphere of a sub-atmospheric mercury pressure of 50–200 microns.

The infiltrant used was the alloy Inconel. The weight of the titanium carbide-nickel skeleton was 79 grams, that of the infiltrant alloy 222.0 grams. The temperature of infiltration was 1450–1480° C. The final density of the infiltrated bucket was 7.08 g./cc., and the total weight 339.8 grams.

EXAMPLE 7

The procedure as described in Example 6 was changed in so far as the metal mixed in powdered form with the titanium carbide was carbonyl iron, and the infiltrant used was the Stainless Steel Alloy No. 310. The weight of the skeleton from titanium carbide-iron was 115.0 grams, that of the infiltrant alloy 209.5 grams. The temperature of infiltration was 1490–1520° C. The final density of the infiltrated, infiltrant-enveloped bucket was 6.72 g./cc., and the total weight 324.5 grams.

It is apparent that variations may be made of the above disclosed invention without departing from the spirit thereof and within the scope of the appended claims.

What I claim is:

1. A composite material shaped blade for turbines in jet engines and the like having a leading edge, a thin trailing edge, and attachment means, said blade being the product of a porous skeleton body of sintered titanium carbide particles, said titanium carbide being present in the skeleton body in amounts ranging from about 50% to about 70% by volume of the skeleton body, substantially the balance of said skeleton body comprising an intercommunicating pore system therethrough containing a ductile heat and corrosion resistant infiltrant metal alloy selected from the group consisting of high temperature cobalt-base, nickel-base and iron-base alloys in amounts ranging from about 50% to about 30% by volume of said skeleton body, said blade having a continuous protective curved surface layer of said infiltrant metal

of a thickness not exceeding ten-thousandths of an inch on the external face thereof, said layer of infiltrant metal being integrally fused to the infiltrated titanium carbide blade body and being thickened at the blade support attaching portion thereof, and being so proportioned at the thin trailing edge of said blade that for thicknesses of about one-thirty second of an inch and less at the extreme edge of the thin trailing edge the material thereof is composed substantially of said infiltrant metal, whereby the trailing edge of said blade is characterized by having increased ductility, improved resistance to thermal and mechanical shock and to bending and fatigue stresses under operating conditions.

2. A composite blade as in claim 1, wherein the thickness of the curved surface layer of infiltrant metal ranges from about one-half of a thousandth to ten thousandths of an inch.

3. A composite blade as in claim 1, wherein the infiltrant metal is a ductile heat and corrosion resistant nickel-base alloy.

4. A composite blade as in claim 1, wherein the infiltrant metal is a ductile heat and corrosion resistant cobalt-base alloy.

5. A composite blade as in claim 1, wherein the infiltrant metal is a ductile heat and corrosion resistant iron-base alloy.

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