United States Patent

Hale

[54] **METHOD OF MAKING A HIGH-SPEED TOOL STEEL**

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[57] **ABSTRACT**

High-speed tool steels are prepared by a powder metallurgical process comprising pressing to shape a powdered mixture of 15–75 weight percent tungsten monocarbide and a matrix of cobalt and iron, the cobalt-to-iron ratio ranging from 0.65 to 2.0. The pressed mixture is then sintered to full density by partially liquefying the ferrous phase of the alloy. The sintered compact is then hardened by heat treatment to transform the ferrous matrix to martensite.

3 Claims, No Drawings

METHOD OF MAKING A HIGH-SPEED TOOL STEEL

This invention relates to a process for the preparation of a shaped, high-speed tool steel alloy of tungsten monocarbide, cobalt and iron.

The quality of high-speed tool steels depends to a considerable extent upon their microstructure, a microstructure normally obtained by hot-working. Thus, high-speed tool steels are processed to the final desired shape by the relatively costly and laborious steps of melting, casting, hot-working (e.g. forging, extruding or rolling) and machining. It would be much more economical to directly cast the desired shape but the as-cast microstructure is very coarse and contains continuous carbide networks which cause severe embrittlement of the casting. The hot-working procedure refines the cast structure to one which is fine and discontinuous and thus much harder and tougher.

It is possible to sinter a fine powder made of a high-speed steel composition to the required full density if a sintering temperature high enough to produce a small amount of liquid phase is used, but when the liquid phase solidifies it forms a continuous brittle carbide network similar to that which occurs in a casting, rendering the sintered part unusable.

Tool steel compositions consisting of Fe—Co—W and C are known. A typical composition of this type contains by weight 40 percent Co, 20 percent W, 1.3 percent C, balance Fe. These steels have been prepared by a modification of the normal process for preparing tool steels, consisting of atomizing a melt into powder, followed by the usual hot-working consolidation steps. The resulting steels exhibit metal-cutting characteristics superior to conventional tool steels. The processing method employed is at least as tedious and costly, however, as that used to make conventional high-speed tool steels.

It is an object of this invention to provide a simpler, less expensive process for making shaped, high-speed tool steels from Fe—Co—W—C alloys.

I have discovered that fine powders of Fe, Co and tungsten monocarbide having from 15–75 percent WC and a weight ratio of Co to Fe between 0.65–2.0 can be pressed and sintered to full density using the technique of producing a partial liquid phase and that the as-sintered structure so produced is free from the usual embrittling network of carbides that forms with other tool steel compositions. The shaped compact is capable of hardening by heat treatment to transform the cobalt-iron matrix to martensite.

This result was quite unexpected since in both the case of conventional high-speed steel compositions, as well as the above-described Fe—Co—W—C compositions, the final structure consists of carbide particles in a martensitic ferrous matrix. All previous experience with such structures indicated that, when sintered to full density without the aid of hot-working, the result was a brittle part. The difference is evidently due to the nature of the carbide phases formed, which turn out to be unique for the Fe—Co—W—C alloys in that the carbides which form in these alloys are essentially entirely of the MC type, whereas the carbides in conventional tool steels are predominantly M_6C and $M_{23}C_6$ types. It is evident that a liquid alloy containing dissolved MC precipitates the MC during cooling as discrete, discontinuous particles while a liquid containing dissolved M_6C and $M_{23}C_6$ precipitates these carbides in the form of continuous brittle networks.

High-speed tool steels are normally characterized as alloy steels capable of being hardened and tempered to a condition of high strength, wear resistance, and hot hardness. In addition, they also exhibit certain metal-cutting performance characteristics which distinguish them from other tool steels and from cemented carbides. The process used in this invention differs from that used to make conventional cemented carbide alloys in several significant respects discussed below. However, the processes contain certain steps in common in that they both ball-mill fine powders, press to shape, and sinter to full density. The compositions used by this process, however, more closely resemble high-speed tool steels in a number of respects. The present alloys have a structure that, except in

the upper portion of the WC range, primarily consists of a ferrous alloy, with carbide particles present as a secondary constituent, while the reverse is true with cemented carbides in that the carbide particles are the predominant phase and an iron-group metal is secondary.

Moreover, the present alloys contain matrices which are hardenable by heat treatment and can be tempered to a condition of hot-hardness, both of which are not characteristic of the matrices of cemented carbides. The liquid-phase sintering of cemented carbides consists of raising to a temperature where essentially all of the iron-group metal becomes molten and the compact is held together by the tightly packed carbide particles. If this same technique were employed with these new tool steel alloys, the compact would melt and completely lose the desired shape. It is essential with these alloys that only a partial liquefaction of the binder phase occur. The sintering temperature must be controlled within a range where a sufficient amount of the ferrous phase liquefies to permit full densification but the amount of liquid is insufficient for the compact to lose shape by sagging or melting. The sintering temperature will vary with the specific composition and with the type of sintering cycle used, but will fall between about 1,200° C. and 1,400° C. This temperature may, however, easily be obtained by heating to a temperature within the range 1,200°–1,400° C. to select, on the one hand, the lowest temperature which produces a product of full density and, on the other hand, to select the temperature above which the pressed mixture begins to sag. Any temperature between these extremes may be used. Normally the range between minimum and maximum will be about 30° C. In the case of a typical composition within the scope of the present invention, 20 percent Fe, 30 percent Co, 46.9 percent W, 3.1 percent C, the critical sintering range is about 1,325°–1,355° C. when the composition is first presintered in hydrogen at 500° C. and then sintered in vacuum.

Heat treatment of the alloys comprises heating to a temperature at which the matrix phase is transformed to an austenitic structure, normally from about 1,000° C. to about 1,200° C. for a period of from several minutes to several hours or even more. The alloy is then rapidly cooled such as by oil quenching to convert the austenitic matrix to martensite. The alloy may then be further hardened by tempering by heating to about 425°–650° C. The tempering step forms a very fine precipitate which increases the hot hardness of the alloy. Both the hardening and tempering treatments are similar to these well-known heat treatment steps with high-speed steels.

I have additionally found that it is possible to further enhance the high temperature hardness of the Fe—Co—W—C tool steel system by use of minor additions of molybdenum, vanadium and chromium, i.e., amounts less than 10 percent by weight of the matrix. When these additions are properly made, the basic structure and heat-treating response are unaltered but a significant increase in hardness is observed at all temperatures up to at least 1,300° F.

The following examples illustrate the practice of the present invention. All parts and percentages are by weight.

EXAMPLE I

An Fe—Co—W—C tool steel alloy of composition 20 percent Fe, 30 percent Co, 47 percent W, 3 percent C was prepared by ball-milling fine powders of Fe, Co and WC for 24 hours in a 4-inch diameter ball mill using acetone as a milling fluid. One percent paraffin was then added as a pressing lubricant and compacts were pressed at a pressure of 30,000 p.s.i. The compacts were then presintered at 500° C. in a hydrogen atmosphere to remove the paraffin and sintered 10 minutes at 1,340° C. in a vacuum furnace during which only partial liquefaction of the binder occurred. The sintered compact was fully dense, as evidenced by the complete lack of porosity in the microstructure, and the microstructure consisted of fine, discrete particles of WC embedded in a ferrous matrix.

The sintered compact was then given a heat treatment consisting of holding 20 minutes at 1,200° C., oil quenching, then tempering 2 hours at 550° C., cooling to room temperature, and another 2 hours' temper at 550° C. After heat treatment the hardness was R_C 70 and the transverse rupture strength averaged 475,000 p.s.i.

Hot-hardness tests showed this material as heat-treated above to be harder than a commercial T–15 tool steel up to 600° C. and slightly softer than T–15 at 700° C. T–15 is a premium high-speed tool steel, generally considered to have best metal-turning properties in the tool steel class. Single-point turning of 1045 steel showed the present alloy to be nearly comparable to T–15 in wear resistance and of essentially similar performance response, i.e., the tool lifeline slope was that of a tool steel rather than a cemented carbide. The performance was thus considered to be very good for a high-speed tool steel prepared by the relatively simple and inexpensive powder metallurgy route.

EXAMPLE II

A basic Fe—Co—W—C tool steel alloy composition was modified by the addition of Mo and Cr to obtain the composition: 23.3 percent Fe, 23.3 percent Co, 47 percent W, 1.5 percent Mo, 1.5 percent Cr, 3.4 percent C. This composition was prepared by the powder metallurgy and heat treatment process as in Example I above. The as-heat-treated hardness

was R_C 74 and the transverse rupture strength was 255,000 p.s.i. The hot hardness of this composition was superior to that of commercial T–15 steel at all temperatures up to 1,300° F. When used to machine 1045 steel as in Example I above, the tool lifeline of this composition was identical with that of T–15 tool steel.

I claim:

1. A process for preparing a shaped, high-speed tool steel alloy comprising

pressing to shape a mixture consisting essentially of 15–75 percent by weight of tungsten monocarbide, and a matrix material, said matrix material comprising cobalt and iron, the cobalt-to-iron weight ratio ranging from 0.65 to 2.0,

sintering said mixture to a fully dense compact at a temperature between 1,200°–1,400° C. to form a partial liquid phase of the cobalt-iron matrix,

and hardening said shaped compact by heat treatment to transform the cobalt-iron matrix to martensite.

2. The process of claim 1 in which the sintered composition is hardened by heating to a temperature at which the iron-cobalt matrix becomes austenitic and then rapidly cooling to room temperature to transform the matrix to martensite.

3. The process of claim 2 in which the sintered composition is further heat-treated by the additional step of tempering by heating to achieve precipitation hardening.

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