Solidification of high speed steels

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High speed steels are ferrous based alloys of the Fe–C–X multicomponent system where X represents a group of alloying elements comprising mainly Cr, W or Mo, V, and Co. These steels are mainly used for cutting tools, since they are characterised by their capacity to retain a high level of hardness while cutting metals at high speed. The as cast microstructure of high speed steels consists of dendrites surrounded by a more or less continuous interdendritic network of eutectic carbides. These are observed even under rapid cooling, at rates as high as 10^6 K s^{-1}. The main features of the as cast microstructure are the distribution and morphology of eutectic carbides, owing to their decisive influence on mechanical properties, and on the service performance of the high speed steels, even after hot working. Hence, the most important improvements on the as cast microstructure that have been obtained through progress in alloy design concern the type, morphology, and volume fraction of the eutectic carbides, while improvements obtained by progress in solidification processing are, on the other hand, mostly related to a significant reduction in the microstructure scale. The development of the microstructure in high speed steels through solidification is reviewed, emphasising the effects of the alloy chemical composition and of the cooling rate. The formation of the eutectic carbides and the techniques used to control their morphology and distribution, both under normal and rapid cooling, are discussed. Microstructures obtained through surface remelting and surface alloying using laser and electron beams are described, and the application of highly alloyed high speed steels to the manufacture of cast hot rolling mill rolls is reviewed. IMR/363

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Introduction

High speed steels comprise a family of alloys mainly used for cutting tools. Their name – high speed steel – is a synthesis of the following two features: (a) the alloys belong to the Fe–C–X multicomponent system, where X represents a group of alloying elements in which Cr, W or Mo, V, and Co are the principal ones; (b) the alloys are characterised by their capacity to retain a high level of hardness even when submitted to elevated temperatures resulting from cutting metals at high speed.[1,2]

A steel developed by Mushet, in England, in the latter half of the 19th century – Fe–2C–2·5Mn–7W (wt-%)* – is considered the forerunner of the modern high speed steels. That steel presented a great advantage over high carbon steels, which had been largely applied so far in cutting tools, since it became hard when air cooled from a temperature from which most steels required water quenching for hardening.[2–4] The most important development during the following 30 years was the substitution of chromium for manganese, as a result of work conducted in Western Europe and the USA.[3]

The full capabilities of the alloy as cutting material were however realised only after the work of Taylor and White on its heat treatment, towards the end of the 19th century. Hardening the alloy from a temperature close to its solidus temperature, they attained uncommonly high hardness levels. Roberts[3] emphasised that, since high speed steel must combine wear resistance with the essential property of red hardness, the first milestone of the high speed steels' history should not be the date when Mushet's steel composition was first proposed, but the date when that composition was combined with this particular heat treatment.

The first alloy formally classified as high speed steel, patented by Crucible Steel Co. in the beginning of this century, contains 0·7C, 18W, 4Cr, 1V (AISI T1 in the American designation terminology or S 18–0–1 in the German designation terminology). During the following 40 years this basic alloy design was held, the most significant improvements being the addition of up to 12%Co and the use of higher carbon and vanadium contents (1·5%C and 4%V, respectively), the latter being partially substituted for tungsten.[3]

During World War II, a general shortage of metallic raw material induced the development of alloy designs with a view to leaner alloy contents and cheaper alloying elements, mainly substituting for tungsten and vanadium.[4] This brought, in the 1950s, a high popularity to AISI M2 steel (S 6–5–2) 0·9C, 6W, 4Cr, 5Mo, 2V, after some restrictions to its use from European steelmakers were surmounted, leading to the overtaking of T1 steel.[5] In the USA, on the other hand, molybdenum rich grades, such as AISI M1 steel (S 1–8–1) 0·8C, 1·5W, 4Cr, 8Mo, 1V, have been used since the 1930s.[4,6] For studying purposes, conventional high speed steel grades have been divided into three series called tungsten, tungsten–molybdenum, and molybdenum high speed steels, according to their W and Mo contents.

In the 1970s, another wave of cost savings based on the use of niobium as an alloying element was attempted.[4,7–9] The main research line was based in the idea of partially substituting niobium, a cheaper and stronger primary carbide forming element, for vanadium in the grades of the tungsten–molybdenum and molybdenum series, thus allowing the vanadium to be utilised at a lower level, predominantly for secondary precipitation hardening. Although the research effort on niobium as a tool steel alloying

* Compositions are given in weight-% unless otherwise stated.

 ISSN 0950–6608

element was developed world wide, the production of conventional niobium alloyed high speed steel grades is currently restricted to South America and the former Soviet Union countries.[10,11]

Despite the diversity of compositions resulting from the alloy design development, the solidified microstructure of high speed steels has maintained its characteristic basic feature, i.e. dendrites surrounded by a more or less continuous interdendritic network of eutectic carbides. These microstructures have evolved over time, however, mainly as a result of changes in two main parameters:

1. Alloy composition. The most important changes caused by progress in alloy design concern the type, morphology, and volume fraction of the eutectic carbides.

2. Solidification process. When considering the role of conventional solidification processing (cooling rate ranging from 10^{-3} to 10^2 K s^{-1}) on the solidified microstructure, the changes in microstructure caused by changing cooling rate are mainly with regard to dendrite size and eutectic colony distribution, both being more homogeneous as a result of the application of electroslag remelting[2,12–14] or refining treatments through minor additions.[15–17] With rapid solidification processing (cooling rate ranging from 10^2 to 10^7 K s^{-1}) the resultant microstructure is mainly characterised by its very reduced scale, the dendrite arm spacing or cell size ranging from 0·01 to 1 μm, i.e. two to four orders of magnitude smaller than that of normal solidified microstructures.[2] Despite this remarkable difference in size, the more or less continuous interdendritic (or intercellular) network of eutectic carbides remains as a distinctive characteristic of the microstructure, even for cooling rates as high as 10^6 K s^{-1}, since short range segregation still takes place. For extremely high cooling rates of more than 10^8 K s^{-1}, however, segregation is suppressed and a featureless microstructure is observed, probably as a result of diffusionless solidification.[18]

The application of rapid solidification techniques, like powder metallurgy, spray deposition, and laser remelting/alloying, allows as a consequence production of conventional high speed steel grades with improved end user properties,[19] as well as design of highly alloyed high speed steel grades with carbide contents up to 50 vol.-%,[18,20] thus ensuring superior wear resistance.[21,22]

Table 1 gives the nominal chemical composition of the most commonly used commercial high speed steel grades, including both the conventional grades (produced through processing routes involving conventional solidification) and the P/M grades (produced through powder metallurgy routes). Figure 1 is a panoramic sketch showing the evolution of the as cast microstructure throughout the century of development of high speed steels.

Solidification sequence of high speed steels

Most of the early experimental work on the solidification path of high speed steels was based on the 18%W, 4%Cr section of the quaternary Fe–W–Cr–C diagram published by Murakami and Hata[23] in the 1930s and modified by Kuo in the 1950s.[24] This particular section received a lot of attention, as it has the same W and Cr contents as AISI T1 steel, largely applied so far as discussed above, Fig. 2.

The following incorrect assumptions were made by early users of the Murakami–Hata–Kuo temperature–concentration section (isoplet) for the analysis of the solidification of high speed steels:

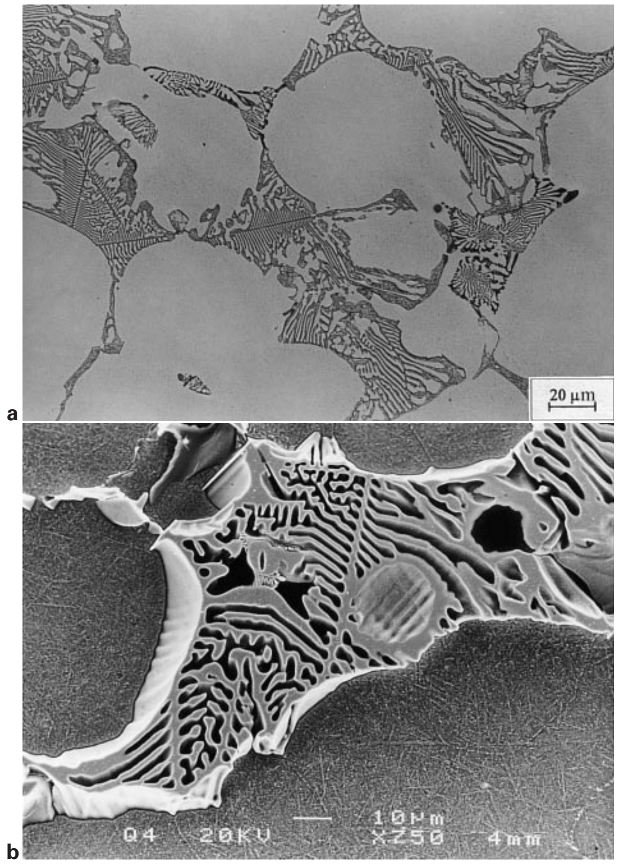
1. The presence of up to 2 wt-%V causes only small changes in the general shape of the section, apart from a slight shift of the whole diagram to higher C contents caused by the stoichiometric consumption of C by the formation of VC carbide, thermodynamically favoured over the W and Cr carbide (M_6C).[24] This assumption is based on the hypothesis that the VC carbide precipitates directly from the liquid, preceding all other solidification reactions. This was soon reviewed by Hoyle and Ineson,[25] showing that vanadium carbides form at the end of solidification, through the eutectic reaction at temperatures lower than that of the formation of M_6C carbide. Additionally it was found that there is a substantial solubility of vanadium in the matrix as well as in M_6C carbides, i.e. not all V is consumed by the formation of MC carbides.

2. Tungsten and molybdenum have almost identical behaviour as alloying elements, as can be seen by the similarity between Fe–W–C and Fe–Mo–C ternary diagrams.[26] On a weight fraction basis this can be expressed by a ‘tungsten equivalent’ value given by Mo/W = 1:2. Actually, that similarity is very limited, since, for the composition ranges typical for high speed steels, the Fe–W–C diagram presents only M_6C carbide, while the Fe–Mo–C diagram presents also Mo_2C.[27]

Table 1 Nominal chemical composition of the most commonly used commercial high speed steel grades

			Chemical composition, wt-%					
Alloy	Series	Processing†	C	Cr	W	Mo	V	Co
T1	Tungsten	C	0·7	4	18	...	1	...
T15		C; PM	1·5	4	12	...	5	5
M1	Molybdenum	C	0·8	4	1·5	8	1	...
M7		C	1	4	1·75	8·75	2	...
M42		C; PM	1·1	3·75	1·5	9·5	1·15	8
M2	Tungsten–	C; PM	1·0	4	6	5	2	...
M35	molybdenum	C; PM	0·8 (1·0)	4	6	5	2	5
ASP 60*		PM	2·3	4	6·5	7	6·5	10·5
CPM Rex 45*		PM	1·3	4	6·25	5	3	8

* Proprietary alloys.
† C produced through processing routes involving conventional solidification; PM produced through powder metallurgy routes.



a optical: etching KOH + $K_3Fe(CN)_6$; *b* SEM

10 Typical morphology of M_6C eutectic in as cast microstructure of M2 steels[45]

M_6C carbide, from which arises secondary platelets of M_6C separated from each other by austenite. These secondary platelets are usually thicker at the end, interrupting the continuity of the austenite and forming a 'wall' of carbide around the eutectic colony.[33,49] Optical microscopy reveals that the cross-sections of the platelets are lamella-like (a 'fishbone' morphology). Figure 10 shows the typical morphology of the M_6C eutectic as seen in optical and scanning electron microscopes. The morphology of the M_6C eutectic is not influenced by chemical composition nor cooling rate (considering values as high as 10^6 K s^{-1}), except that at faster cooling rates the distance between platelets is decreased.[33,49]

In the chemical formula for the M_6C carbide, the letter M corresponds to the elements W, Mo, V, Cr, and, mainly, Fe. The elements W and Mo are dissolved in the same proportion as they are present in the composition of the steel, which is not the case for V, whose low level of dissolution is a characteristic of the M_6C carbide.[33] The M_6C carbide has a complex fcc crystalline structure and its hardness is around 1500 HV.[50]

M_2C eutectic

Contrary to what occurs with M_6C eutectic, the morphology of M_2C eutectic is substantially influenced by various factors, namely, the chemical composition,[29,31,49,51] the cooling rate,[29,31,51,52] and minor additions of certain elements such as aluminium and nitrogen.[53–55]

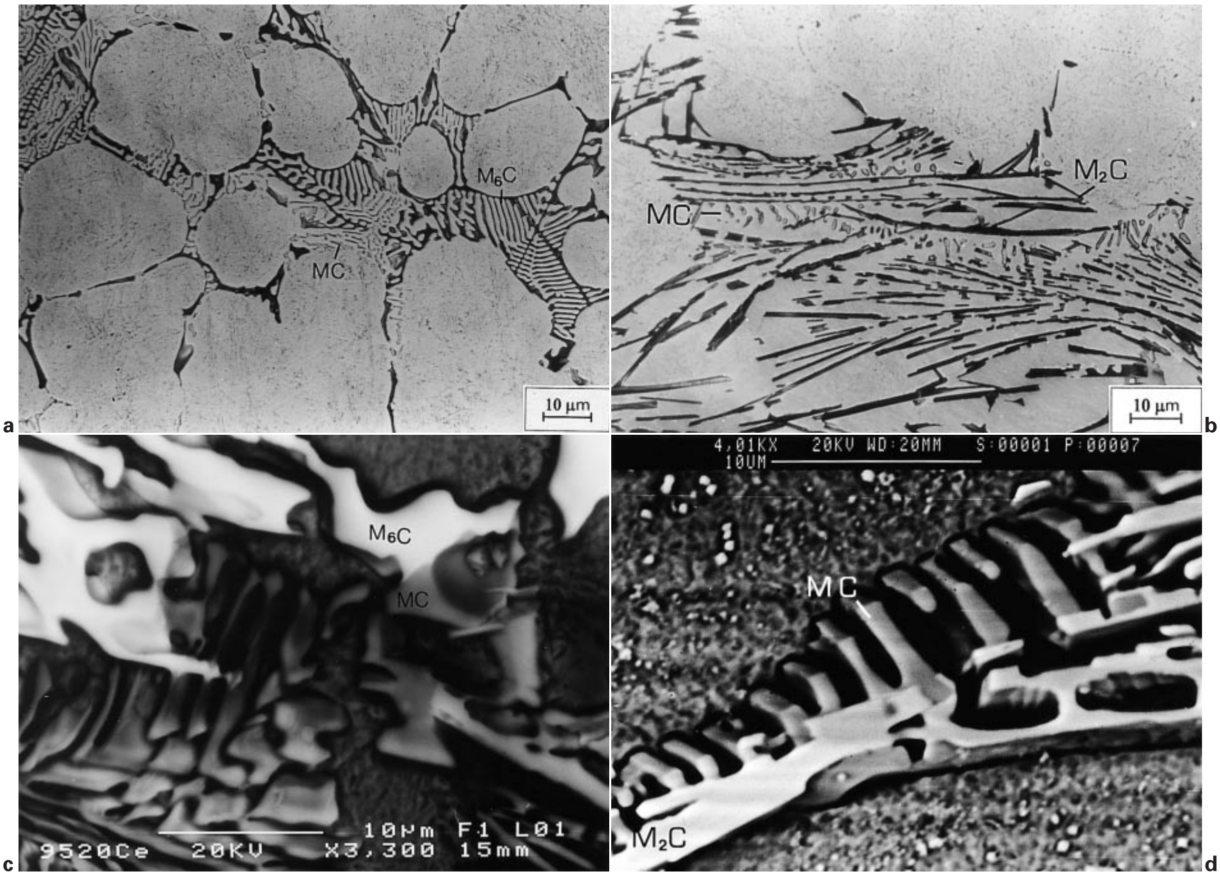
A new terminology was proposed for the classification of the morphological types of the M_2C eutectic,[56] using the formal classification for all eutectic systems proposed by Crocker *et al.*[57] This was necessary to describe the characteristics of this eutectic more precisely and to avoid conflicting uses of the same expression, as commonly observed in the literature.[56] Figure 11 shows these morphologies and related classification.

Strictly speaking, there are two types of M_2C eutectic morphology: irregular and complex regular. Irregular M_2C eutectic is characterised by a ragged boundary that does not clearly outline the interface between the matrix and the eutectic pool, suggesting a limited coupling between eutectic phases. In this eutectic, M_2C carbide has a platelike morphology and a tendency to assemble as radiating clusters (this feature is often referred to as 'fanlike' or 'feathery'). It also shows extensive branching, which is responsible for avoiding overgrowth by austenite and, consequently, for maintaining M_2C carbide as the leading phase.[45] Complex-regular M_2C eutectic is characterised by a smooth boundary that clearly outlines the interface between the matrix and the eutectic pool; it shows a regular distribution of M_2C eutectic carbide over small areas, forming cells with macrofacets.[56] It was shown that complex-regular M_2C eutectic grows as a spiral eutectic,[58] though perfectly developed spirals are rarely observed in the microstructure of as cast high speed steel. Instead of this, degenerate or interrupted spirals characterise the complex-regular M_2C eutectic, mainly due to the low volume fraction of eutectic, as well as to the highly branched dendritic structure in high speed steel.[45]

Irregular M_2C eutectic is chiefly promoted by low cooling rate or high vanadium content and the complex-regular M_2C eutectic by high cooling rate or low vanadium content.[29,31,45,52] The effect of higher vanadium content is to decrease the volume fraction of M_2C carbide in M_2C eutectic,[45] thus increasing the difficulty of keeping M_2C carbide as the leading phase in the eutectic, and an irregular structure then arises. On the other hand, as the cooling rate increases, the difference between the growth rates of M_2C carbide (faceted phase) and austenite (non-faceted phase) also increases and the eutectic morphology becomes strictly controlled by the characteristics of the M_2C carbide growth (hexagonal hopper crystals[45,58]), ensuring the formation of a complex-regular structure.

A third type of morphology associated with faster cooling rates and characterised by the presence of rod-shaped M_2C carbide was identified by Fredriksson and Nica[31] and Lichtenegger *et al.*[59] Fredriksson and Nica verified that this morphology predominated in the regions of the samples whose directional solidification was concluded by quenching. It is believed that this morphological type just corresponds to a 'broken complex-regular' morphology that comes from frequent interruptions of the spiral growth of the M_2C carbide caused by localised overgrowth by austenite, leading to its rod-shaped appearance.[45]

The M_2C carbide dissolves all the principal elements of the high speed steels. Moreover, the composition of M_2C carbide presents great variability



a, *b* optical, etching KOH + $K_3Fe(CN)_6$; *c*, *d* SEM

20 *a*, *c* duplex M_6C–MC eutectic and *b*, *d* duplex M_2C–MC eutectic in rare earth modified M2 steel[45]

hot pressing, or extrusion and sintering, injection moulding and sintering, and hot isostatic pressing. The products obtained by these routes exhibit superior properties and performance to those manufactured by the classic route (i.e. by normal cooling), mainly with respect to toughness, isotropy of properties, size stability in heat treating and grindability,[70,88,89] owing to their fine scale microstructures.[2,19,87] Moreover, the refinement of the eutectic carbides makes it possible to produce high speed steels with higher alloying content, mainly carbon and vanadium, thus ensuring better wear resistance of the products.[18,86]

The cooling rate during the solidification of gas atomised high speed steel particles in the size range below 500 µm is estimated to fall in the range 10^4–10^6 K s^{-1} (Ref. 87) – the smaller the particle size the higher the cooling rate. Moreover, the solidification of splat caps encasing some particles, formed as a result of in-flight collision between completely liquid droplets and fully solidified particles, is estimated to occur at cooling rates exceeding 10^8 K s^{-1} (Refs. 87, 90). Since the resulting powder consists of particles of widely varying sizes, and since collisions are a common feature of the process, a spectrum of size scale of the microstructure can be observed in a single atomised charge (Fig. 21).[91] The as solidified microstructure of the atomised particles is two to three orders of magnitude more refined than those obtained for conventionally cast ingots.

Depending on the particle size, the morphology of the microstructure of the high speed steel powder particles is dendritic, cellular, or a mixture of both, the fraction of cellular morphology being significant only in the very small sized particles, i.e. those solidified at higher cooling rates.[92,93] The morphology changes to a globular one in the Nb-containing high speed steels, which is explained by the presence of primary NbC acting as nucleation sites for the primary solid solution phase.[94]

The existence of undercoolings high enough to promote the formation of a cellular morphology is attributed to the higher cooling rate and to the lesser number of potential heterogeneous nucleation sites for solidification of the particles.[92] Figure 22 shows micrographs of particles with dendritic and cellular morphologies, as well as the relationship between the size of the particle and the fraction of the particles with cellular structure for a high speed steel (Fe–1·2C–4Cr–6Mo–4W–1V–1Nb–12Co).

Except for its smaller scale, the solidified microstructure of the high speed steel powders is not significantly different from that of the slow cooled high speed steel ingots, since it is unlikely that the particles are sufficiently undercooled to allow diffusionless solidification, owing to the large freezing range of those steels.[95] Thus, the normal microsegregated structure constituted by solid solution dendrites (or cells) delineated by a more or less continuous interdendritic (or intercellular) network of