Vanadium Microalloyed Forging Steel

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Vanadium microalloyed steel forgings provide lightweight, high performance and cost effective solutions for many critical automotive components in the drivetrain and chassis of modern cars and trucks. These high strength steel forgings are already widely used in regions with developed automotive industries such as Europe, Japan and North America, and they play an important role in new vehicle designs required to meet increasingly stringent weight and emission targets. Developing countries and regions will also present a significant opportunity for adoption of this technology as indigenous automotive production expands and the best engineering solutions are applied. The majority of these forgings are produced from medium carbon steels with small additions of vanadium to provide substantial precipitation strengthening of the controlled cooled ferrite-pearlite microstructure. The aim of this paper is to provide an introduction to the essential aspects of this important technology by providing an overview of the historical development, basic metallurgy, processing, properties and typical applications.
Keywords: vanadium, forging steel, automotive, fatigue, machinability

Introduction

Air cooled medium carbon vanadium microalloyed (MA) ferrite-pearlite forging steels have been used in a wide range of automotive components with ultimate tensile strength up to 1100 MPa as a replacement for the traditional quenched and tempered (QT) steels, particularly in Western Europe, USA and Japan since the introduction of the first grade (49MnVS3) in Germany in the mid-1970's. The advantages of significant cost saving due to elimination of post forging heat treatment, straightening and stress relieving, reduced distortion, improved machinability, more consistent properties and lower material cost compared to QT alloy grades, as well as higher productivity are the driving forces for the application of these steels. By appropriate alloy design and processing, vanadium MA forging steels have attained the strength levels equivalent to QT steels [1,2,3]. The fatigue resistance of vanadium MA forging steels is comparable with that of QT steels at the same hardness level. However, the toughness at the same strength level is inferior to that of QT steels but adequate for many applications, such as connecting rods and crankshafts, which are subjected to cyclic fatigue loading rather than impact loading in service [2,4]. However, for safety critical components in the steering and suspension systems, a higher level of toughness is sometimes considered desirable.

MA forging steels achieve the desired high strength through precipitation strengthening. Because of the high solubility of vanadium in austenite, it is preferred over niobium or titanium as the microalloying additive in medium-carbon steels (figure 1). Vanadium can be fully dissolved in the austenite during normal reheating and precipitates as fine V(C,N) particles in proeutectoid ferrite as well as in ferrite lamellae of pearlite on cooling after hot forging, which provides significant increase in strength regardless of carbon content [5,6]. Further increases in strength can be achieved by increasing the nitrogen content up to 200 ppm, or higher. In addition, the high solubility of V(C,N) in the austenite permits the use of lower reheating temperature, which means lower forging production costs. Furthermore, microstructure and mechanical properties of vanadium MA steels don't vary much with changes in reheating temperature, a benefit to produce forgings with uniform and consistent properties. Vanadium MA forging steels exhibit similar induction hardening characteristics to QT steel with equivalent carbon content and respond well to nitriding, carbonitriding and nitrocarburing [7].

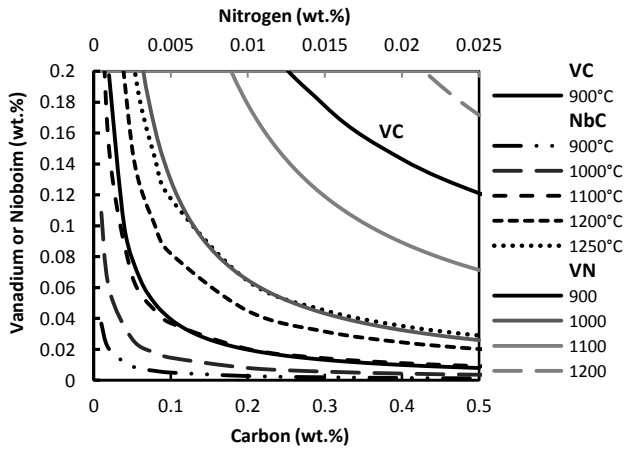

Fig. 1 Solubility isotherms of VC, VN and NbC in austenite and the solubility products used to calculated the diagrams are according to Tukdogan [8]

Alloy Design

The European standardized MA forging steel grades are given in Table 1 [9] and these are representative of the types of steels used globally. The majority of the steels are vanadium MA medium carbon steels (0.3 - 0.5%C) and the composition is selected depending on the mechanical properties required.

Table 1. Chemical composition of European standard air cooled MA Forging Steels (wt.%)

Grade	C	Si	Mn	P	S	V	N
19MnVS6	0.15/0.22	0.15/0.80	1.20/1.60	≤0.025	0.020/0.060	0.08/0.20	0.01/0.02
30MnVS6	0.26/0.33	0.15/0.80	1.20/1.60	≤0.025	0.020/0.060	0.08/0.20	0.01/0.02
38MnVS6	0.34/0.41	0.15/0.80	1.20/1.60	≤0.025	0.020/0.060	0.08/0.20	0.01/0.02
46MnVS6	0.42/0.49	0.15/0.80	1.20/1.60	≤0.025	0.020/0.060	0.08/0.20	0.01/0.02
46MnVS3	0.42/0.49	0.15/0.80	0.60/1.00	≤0.025	0.020/0.060	0.08/0.20	0.01/0.02

Carbon is the most effective element for strengthening these steels and increasing carbon content increases pearlite volume fraction, resulting in higher strength, but lower toughness. The high level of manganese content is used to provide solid solution strengthening and reduces the transformation temperature, enhancing strength. Carbon and manganese levels can be optimised to control the transformation temperature range, which provides an improved combination of strength, toughness and ductility. In most MA forging steels, the silicon content is about 0.30%, but up to 0.80% of silicon is also used to increase ferrite content, resulting in improved toughness. Sulphur addition is made for improving machinability and sulphur content is selected depending upon the machinability requirements. A favourable sulphide shade and distribution can be obtained by calcium or tellurium treatment. Increasing sulphur content has also been found to increase the amount of intragranular ferrite, which improves the toughness.

The addition of vanadium provides precipitation strengthening obtained from interphase precipitation of V(C,N) in ferrite and the ferrite lamellae of the pearlite, resulting in an increase in yield and tensile strength. The vanadium level can be selected to produce the desired strength level. Nitrogen (up to 0.02%) may be added to further increase precipitation strengthening.

In some steels, in order to improve ductility and toughness, additions of titanium (0.010-0.015%) are made to control the austenite grain size on reheating. However, it is important to control the Ti:N ratio (<3.42) in order to avoid coarse TiN particles, which are detrimental to toughness.

Processing of Microalloyed Forging Steels

Vanadium MA forging steels can be processed by conventional forging practices with a minimum of process adjustments [3]. The key process parameters which must be controlled are soaking temperature, forging temperature, cooling rate immediately after forging and during the transformation, and binning/stacking temperature.

In general, the soaking temperature can remain unaffected, although for vanadium MA steels, a soaking temperature of 1150°C-1250°C is recommended to ensure solution of all initial V(C,N) precipitates and realise the full strengthening potential of vanadium additions. Close control of the reheating and soaking process is necessary to obtain consistent finish part properties. There is an increase in tensile strength with increasing soaking temperature and this is associated with a coarse austenite grain size on reheating, which influences the final prior austenite grain size and results in a greater proportion of pearlite. However, a too high soaking temperature (exceeding 1250°C) may cause excessive grain coarsening and it should be avoided. The soaking time is not critical and both furnace and induction heating can be used. The soaking temperature is also influential in that it controls the finish forging temperature.

During the hot forging process, after each deformation stage, the austenite is recrystallised and this causes a degree of grain refinement that is enhanced by reducing the forging temperature. The forging temperature range does not greatly affect the strength of vanadium MA steels [10]. Tensile strength decreases slightly with decreasing the finish forging temperature as a result of a reduction in the proportion of pearlite, but there is no significant effect on yield strength. Ductility and toughness show a significant increase with a reduction in finishing temperature and this is due to grain refinement of the austenite and increased ferrite content. Therefore, reducing the finish forging temperature offers an opportunity of improved ductility and toughness. However, lowering of the finish forging temperature can increase the flow stress of the steel, which leads to higher forging loads and increased forming energy and consequently shorter life-span of the forging tools.

The cooling rate after forging is an important parameter for control of the microstructure/properties developed in vanadium MA steels. In general, most forged components can be air cooled on racks or on a conveyor with satisfactory results. However, both strength and toughness are impaired if the forgings are cooled in bins with very slow cooling rate, which results in coarse pearlite microstructure and should be

avoided. Forced air-cooling may be necessary, especially in the case of large forging section sizes which would otherwise cool very slowly. Increasing the cooling rate generally increases the yield and tensile strength, with little or no reduction in toughness due to less austenite grain growth, which results in finer ferrite, a smaller pearlite interlamellar spacing together with more effective precipitation strengthening. However, there is an upper limit to the cooling rate after forging because of the risk of formation of very brittle high carbon bainite, which could result in a significant drop in yield strength and toughness, especially ductility, as well as degraded machinability. The preferred cooling rate suggested ranges from 1-5°C/s [11] and should be applied from the finishing temperature to about 600°C. In addition, the cooling rate should be consistent from part to part within and between batches.

The binning or stacking temperature can influence the properties and should be below 600°C to ensure that the transformation from austenite to ferrite-pearlite has occurred during air cooling and no significant coarsening of V(C,N) particles takes place [1,12].

Microstructure and Properties Relationship

1. Strength

Tensile strength of 900-1100 MPa and yields strength of 420-700 MPa with reasonable ductility can be attained in air cooled vanadium MA medium carbon forging steels, and these strength levels are suitable for many automotive applications, such as connecting rods, crankshafts, steering linkage and stub axle. An example of a typical ferrite-pearlite microstructure observed in these steels is shown in figure 2 [13], and the strength increases with higher volume fraction of pearlite, smaller interlamellar spacing of pearlite and enhanced precipitation strengthening of V(C,N).

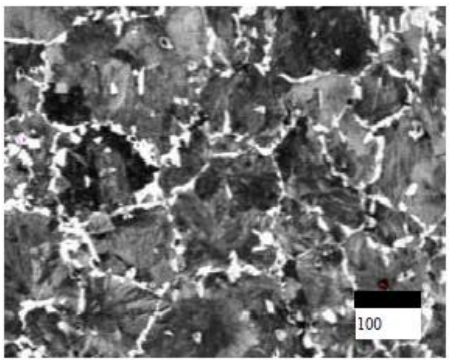

Fig. 2 Typical pearlite+ferrite microstructure of air cooled microalloyed forging steel [13]

The volume fraction of pearlite is mainly a function of carbon content and it increases with higher carbon level. In addition, the volume fraction of pearlite is also influence by the prior austenite grain size, manganese level and the cooling rate following the forging. A coarser prior austenite grain size contributes to a higher volume fraction of pearlite due to less nucleation sites for ferrite. Increasing cooling rate or manganese level decreases the transformation temperature, which results in increased volume fraction of pearlite. The pearlite interlameller spacing is predominantly determined by cooling rate after forging and manganese level. Again, increasing the cooling rate or manganese level decreases the transformation temperature, which provides a higher driving force for nucleation of cementite carbide laths and shorter diffusion distances for the carbon, resulting in decreased interlameller spacing.

Interphase precipitation of V(C,N) in proeutectoid ferrite and ferrite lamellae of pearlite (as shown in figure 3 [14]) provides significant contribution to the strength and this effect is almost independent of carbon content. It has been reported that yield strength and tensile strength increase linearly with increasing vanadium content up to 0.20% and 0.01%V addition contributes about 10 MPa to the tensile strength and about 12 MPa to the yield strength, as shown in figure 4 [15]. A further increase in the strength can also be seen at the higher nitrogen level. Nitrogen in vanadium MA steel increases the driving force for V(C,N) nucleation, which results in larger volume fraction of V(C,N) precipitates, a smaller average particle size and inter-sheet spacing, resulting in higher precipitation strengthening. Increasing cooling rate after forging or manganese level is beneficial for precipitation strengthening since the transformation temperature is decreased by faster cooling rate or higher manganese content and lower transformation leads to a smaller inter-sheet spacing.

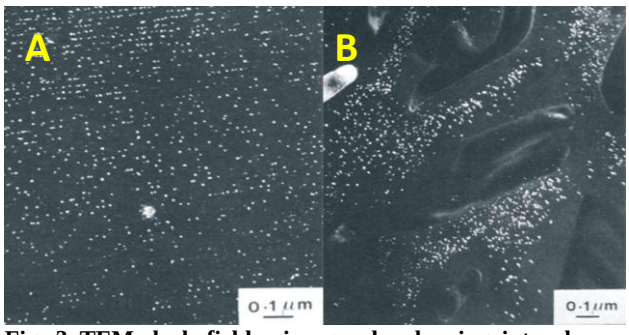

Fig. 3 TEM dark field micrographs showing interphase precipitation of VC in proeutectoid ferrite and ferrite lamellae of pearlite [14]

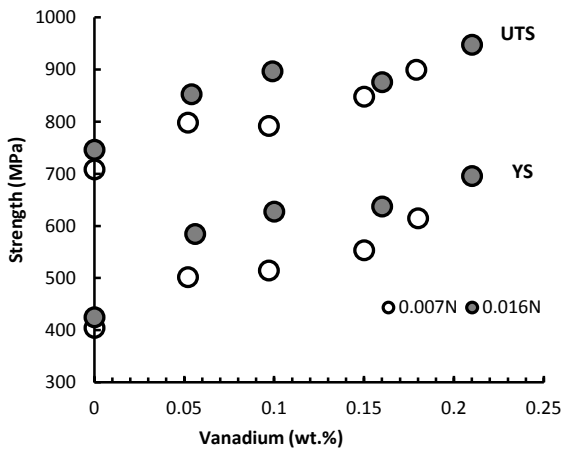

Fig.4 Effects of vanadium and nitrogen on the strength of steels containing 0.41%C and 1.50%Mn [15]

Vanadium MA forging steels exhibit a uniform microstructure and strength from the surface to the centre of the forgings over a wide range of part sizes and shapes [16]. On the other hand QT steels which have a tempered martensite microstructure can show a marked increase in coarseness of microstructure in the centre which can give rise to significant difference in strength between the centre and the outside of the forging. In order to achieve adequate consistency of properties in QT steel forgings, relatively high carbon content and additions of expensive alloying elements including chromium, molybdenum and nickel are necessary to increase the hardenability.

2. Toughness

The toughness levels currently available in MA forging steels are adequate for many applications, such as crankshafts, connecting rods and wheel hubs, which are not subjected to severe impact loading, although the values are lower than that of QT steels at the same strength level. This has been demonstrated in many cases by component testing of automotive parts under field conditions, which show that the impact toughness values, as determined in the notch impact test, are not as important for the life of the component as they are often considered [2,4]. This can be explained by the fact that components do not normally experience such a fast impact speed as in the impact test procedure. In addition, the widespread use of, for example, cast iron crankshafts and forged steel fracture split connecting rods with significant lower ductility and toughness is another indication of the non-criticality of the toughness of these particular parts. Ultimately it is the responsibility of the vehicle or system manufacturer to validate acceptable component performance through appropriate testing procedures.

The lower toughness in air cooled forging steels is attributed to the coarse grained ferrite-pearlite microstructure, and the toughness can be improved by a finer pearlite colony size and thinner cementite lamellaes. The pearlite colony size controls the spread of cleavage fracture and it is primarily determined by the size of prior austenite grain size. The prior austenite grain size can be refined by lowering the soaking temperature and the forging finishing temperature, and increased cooling rate after forging. Unfortunately, lower finish forging temperature is often impractical because it increases forging loads/energy, shortens the die life and possible delays in the hot deformation sequence to finish colder. Most forgers are reluctant to adopt this method to increase toughness [1,6]. Prior austenite grain size refinement can also be achieved by a small addition of titanium (0.010-0.015), which forms extremely stable TiN particles in austenite. These TiN particles can inhibit austenite growth during reheating and limit grain growth after forging, resulting in a finer prior austenite grain size before the transformation. Ti/N ratio is an import factor, which influences the size and distribution of TiN precipitates. It is suggested [17] that the Ti/N ratio should be lower than the stoichiometric ratio of 3.42 to control TiN particle size. However, if too much titanium is added, very high supersaturation results in precipitation of TiN particles at near or above the solidus temperature and those TiN particles are too coarse (1-5 μm) to pin the austenite grain boundaries, but detrimental to toughness, acting as cleavage fracture nucleation sites. In addition, it is proposed that the Ti-route would normally require continuous casting since high cooling rates during the solidification process are needed to produce finely dispersed TiN particles for grain growth control. Deceased cementite lamellae thickness is beneficial because thicker carbides crack at a smaller strain value and voids form more easily. The cementite lamellae thickness is largely controlled by the transformation temperature. Increasing cooling rate after forging or manganese content lowers the transformation temperature, which increases the nucleation rate of pearlite and decreases the growth rate of cementite lamellaes, resulting in a thinner cementite lamellae thickness.

Furthermore, the use of reduced carbon content with a compensating increase in solid solution strengthening by silicon has been effectively exploited, although there appears to be an optimum silicon content (0.6-0.8%), above which the toughness deteriorates [2,18]. It is also suggested that the increased silicon content improves the ferrite distribution and the ferrite is more distributed within

the grains instead of being a network in the grain boundaries.

Improving the toughness by using intragranular nucleation of ferrite has been also reported [19]. The intragranular ferrite transformation sites are identified as V(C, N) or VN precipitated on MnS [20]. V(C,N) has very high lattice coherency with ferrite and reduces the energy required for ferrite transformation and this is the reason why vanadium plays its role as intragranular ferrite nucleation sites.

3. Fatigue Strength

Fatigue strength is generally the dominant engineering performance criterion in forged automotive components because it has been estimated that fatigue fracture contributes to approximately 90% of the mechanical service failures [21]. Fatigue tests on smooth and notched specimens and on actual automotive components by many research groups, steel and automotive companies [13,22,23,24,25,26,27] have demonstrated that vanadium MA forging steels have comparable fatigue properties to QT steels.

Babu and Gromer [28] conducted fatigue tests on both smooth and notched specimens for comparison of fatigue behaviour of the vanadium MA (SAE 1541+V) steel and the QT (SAE 4140) steel. Strain controlled axial fatigue testing of smooth specimens showed that the strain-life fatigue curves for the SAE 1541+V steel and the SAE 4140 are nearly identical. Limited fatigue testing of notched specimens also indicated that the notch fatigue strength of the MA steel is very similar to that of the QT steel and both of the steels exhibited similar notch sensitivity. In addition, they found that the fatigue life of the MA steels follow the same relationship to hardness as the QT steels.

In addition, it has been reported [27,29] that comparison of cyclic and monotonic stress-strain curves for the vanadium MA steels and the QT steels shows a major difference in response to the cyclic loading. The MA steels exhibit significantly less cyclic softening than the QT steels and it is suggested that this is attributed to the fine precipitates of V(C,N), which increase the resistance to dislocation slip. The higher cyclic softening of the QT steels decreases the maximum tolerable stress which a material can sustain during fatigue.

Comparative fatigue tests on components, such as crankshafts, connecting rods, front axles and steering knuckles, made of both of MA and QT steels have proved that the MA forged components have equivalent, or in some cases better fatigue strength compared to QT forged parts in both the low and high cycle life regimes [30]. An example of a comparison of fatigue strength of forged vanadium MA steel (SV40CL1) connecting rod, which was developed for Nissan's high performance engines and QT steel (S40C) connecting rod [26] is given in figure 5. Kuratomi, et al. concluded that the MA steel connecting rod exhibits 25% higher fatigue limit than the similar forged QT steel. In addition, the connecting rods made of the MA steel are 10% lighter in weight and the processing time is reduced by 30%. On the development of high performance crankshafts for four and six cylinder engines at Chrysler Corporation, Hoffmann and Turonek [31] showed that the fatigue strength of two vanadium (0.06-0.08%V) MA grades met or exceeded the fatigue strength of the QT carbon steel (SAE 1050) and QT alloyed steel (SAE 4140) for the applications. They also suggested that the MA grade could reduce the finished cost by 11-19% compared to the QT steel SAE 4140 and by 7-11% compared to the QT steel SAE 1050. All of these results indicate the promising application of MA forging steels for automotive components.

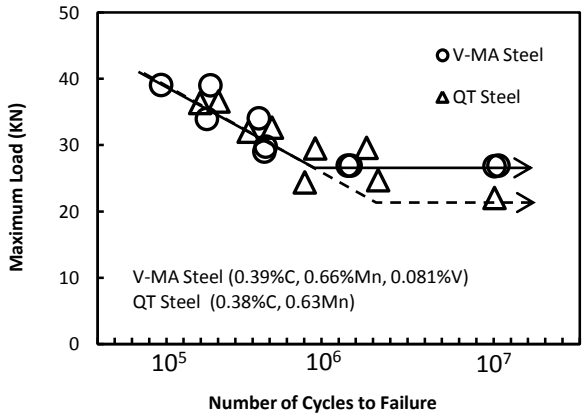

Fig. 5 Fatigue properties of connecting rods [25]

4. Machinability

Machinability is an important property of forging steels since most forged components are machined to final shape and tolerance following forging. It is well recognized that the machinability (based on tool life, cutting force and surface/integrity quality) of vanadium MA forging steels is superior compared to that of QT steels at equivalent hardness levels [32,33] and this is due to more favourable ferrite-pearite microstructure and more consistent hardness of vanadium MA forging steels. The improved machinability of vanadium MA forging steels offers major savings potential, considering machining is the greatest individual cost in the manufacture of a forged component, which can be as much as 50% of the total cost of a part [2].

The superior machinability of vanadium MA forging steels has been demonstrated by many published papers and an example in figure 6 shows that the vanadium MA steel (30MnVS6) exhibits longer tool life in comparison with the QT steels for different cutting speed and specific feed rate [32]. This work has also indicated that at any conditions, the cutting force for the MA steel was lower than the QT steels and suggested that the lower cutting forces for the MA steel arise from improved chip formation. Studies of the chip formation process [32,33] indicated that the MA steels have lower chip thickness leading to smaller contact lengths which in turn result in lower cutting force.

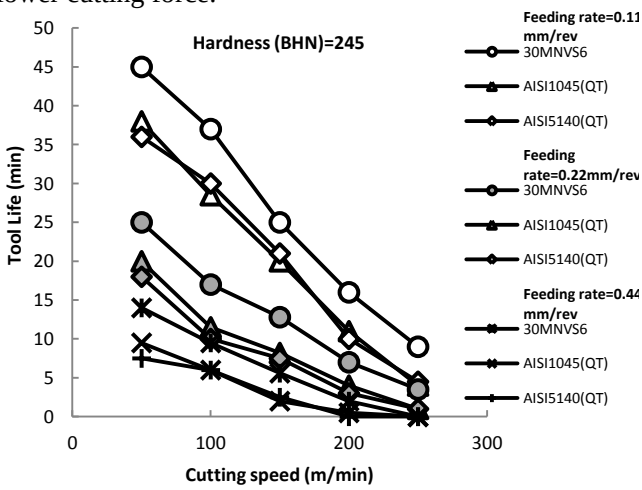

Fig. 6 Tool life at different cutting speeding and feed rate in turning of steels [32]

Bhattacharya [33] suggested that the ferrite-pearlite microstructure of MA steels concentrates the deformation into a thin localized region, while the quenched the tempered martensite microstructure of QT steel, on the other hand, has a larger shear flow zone, which results in a lower chip shear angle, thicker chips and higher forces. In addition, the ferrite-pearlite microstructure causes easier initiation of micro-fractures, and thus limits the flow resistance and this will lead to easier chip formation. Furthermore, observations of the top side of the chips indicate the presence of adhered build-up-edge from the MA steel, while the QT steel shows no indication of any build-up-edge formation. The formation of build-up-edge will also act to both increase the shear angle and thus reduce cutting force and may protect the tool to yield increased tool life.

The machinability of MA forging steels can be further improved by slightly enhanced sulphur levels (up to 0.07%) with optional sulphide shape modifications by calcium and/or tellurium addition to improve transverse ductility and toughness. Addition of sulphur produces MnS particles, which provide lubrication to the cutting tool as well as enhance short-breaking chips. Ideally, MnS particles should have a certain size/shape and be evenly distributed.

Applications

Air cooled vanadium MA forging steels have become increasingly popular in automotive applications including crankshafts, connecting rods, steering knuckles, axle beams and tension rods and they are the “first choice” when selecting materials for new automotive components in Europe. The use of vanadium MA forging steels result in significant cost savings compared to traditional QT steels mainly through elimination of heat treatment, improved machinability and increased productivity. Hoffmann and Turonek [31] examined the cost reduction opportunities associated with two vanadium MA steels compared to medium carbon steel SAE1050 and medium carbon alloy steel SAE4140 grades and these same grades at a higher sulphur level of 0.1% (SAE1050-HS and SAE4140+HS) for high performance four and six cylinder crankshafts and divided the principle cost into raw material, forging, heat treatment and machining. Figure 7 shows the costs at three production levels for six cylinder crankshaft. It can be seen that the material cost of the MA steel is about 2% less than the SAE4140 grade and 10% less than the SAE4140+HS grade. The forging cost is about 7% less for the MA steel. The significant cost saving for the MA steel is machining, especially with increased production volume. At annual volume of 300,000 parts, the machining cost of the MA steel is about 28% lower than the SAE4140 grade and 8% lower than the SAE4140+HS grade. This study concluded that the use of MA steel could reduce the finished cost by 7 to 11 percent compared to the QT carbon steel (SAE 1050) and by 11-19% compared to the QT alloy steel (SAE 4140). An extensive study by Nallicheri et al. [34] on material alternatives for automotive crankshafts, based on manufacturing economics, also indicated that production volume is a predominant factor in cost effective production, and at production volumes above 200,000 parts per year, microalloyed steel forgings offer the most cost effective solution for high performance crankshafts.

These examples demonstrate the economic benefits of using vanadium MA forging steels and illustrate the important factors to be considered when comparing alternative options. These factors are relevant at different stages of the component supply chain, including the steelmaker, forger, machinist and system/vehicle designer/manufacturer, and in order to ensure the successful application of vanadium MA forging steels there must be close co-operation and sharing of economic benefits. The transition to

vanadium MA forgings normally requires a development plan including controlled trials at each stage of the supply chain to acquire essential data and enable the establishment of robust process control plans to ensure reproducible mechanical properties and product/process performance.

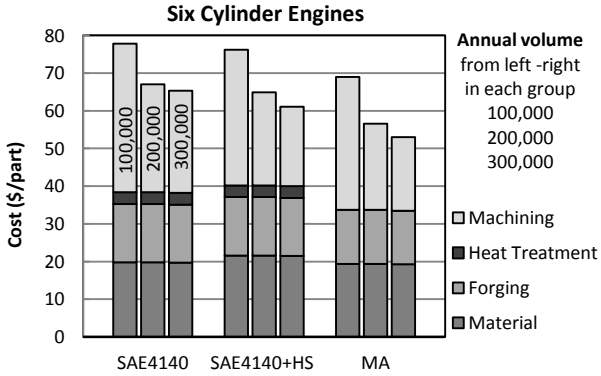

Fig.7 Cost data for six cylinder crankshaft [31]

The development of vanadium MA forging steels is ongoing and there are many new alloys that go beyond the conventional ferrite-perlite grades discussed in this paper and designed to suit specific requirements and with appropriate properties. For example, forged steel fracture splittable connecting rods have been developed with a wide range of chemical compositions to optimise fatigue strength, machinability, splittability and cost, and some of these take advantage of precipitation strengthening effect of vanadium at levels in excess of 0.2%. There are also air cooled bainitic forging steels, used for high strength crankshafts and diesel injection components, that exploit the strengthening effect of vanadium to exceed the performance available from ferrite-perlite grades. We should expect these developments to continue and for vanadium to be used in increasing quantities to achieve the required weight reduction and environmental targets with optimum economic value.

Conclusions

This paper has provided a short summary of some of the more important technical information relating to the design, processing, properties and applications of vanadium MA forging steels.

The main emphasis has been on the most important class of these steels – the widely used medium carbon air-cooled ferrite-pearlite grades.

It has been highlighted that vanadium is the most important and widely used microalloying element in these steels, and that it enables them to provide a cost effective and high fatigue performance alternative to carbon or low alloy quench and tempered steels.

The relatively low impact toughness of these steels, compared with conventional quench and tempered steels, has been considered and it has been suggested that this perceived performance deficiency is of little practical importance in many automotive components, especially within the engine. Furthermore it is suggested that component designers should consider the overall performance criteria and utilise appropriate design and validation methods rather than simply adopting established specification requirements based on QT steels.

The need for supply chain partners to cooperate during the transition process by conducting production trials, developing process control plans and sharing economic benefits has been highlighted.

Brief mention has been made of more recent developments that will continue to exploit and expand the use of vanadium, notably in forged steel fracture split connecting rods and air cooled bainitic crankshafts.

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