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MECHANICAL PROPERTIES AND MACHINABILITY OF A HIGH STRENGTH, MEDIUM CARBON, MICROALLOYED STEEL

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ABSTRACT

The development of a high strength (tensile strength beyond 900 N/mm^2), medium carbon, vanadium microalloyed steel for hot forged automotive components has been reviewed in the paper.

The influence of different alloying elements was investigated. The most effective elements to increase the strength were chromium and manganese. In TEM investigation was found that in comparison with the lower strength melt chromium + manganese alloyed steel showed a high density of small V(C,N) precipitates. Mechanical properties were studied in tensile, impact, fatigue and fracture mechanics tests. Most mechanical properties of the microalloyed and Q&T steel were essentially equal. Exeptions were the lower impact strength and higher fatigue strength of the microalloyed steel. The possibilities to improve the impact strength of the microalloyed steel are being considered. Machinability was tested in different operations. Generally the machinability of the high strength microalloyed steel was comparable with Q&T steels. However, a different behaviour was found in deep hole drilling. The use of calcium treatment to improve the machinability is discussed.

The high strength microalloyed grade developed is able to substitute for alloyed Q&T steels in most components. When weight saving is desired, the possibility of substituting for lower strength microalloyed steels by the high strength grade is considered.

INTRODUCTION

Ferritic-pearlitic microalloyed steels have substituted for Q&T steels in hot forgings due to the reduction in production costs. The principal saving is achieved by the elimination of the separate Q&T treatment after hot forging. There is also a decreased need for cold straightening and stress relieving annealing.

Untill now, microalloyed steel have substituted mainly for the low strength, i.e. unalloyed Q&T steels. However, great potential cost reductions lie in the substitution of the higher strength, alloyed Q&T steels with the tensile strength beyond 900 N/mm^2.

The purpose of this paper is to describe the development, properties, and uses of a new, high strength, microalloyed steel with a minimum tensile strength of 900 N/mm^2.

LABORATORY SCALE DEVELOPMENT

In the laboratory scale experiments the influence of the chemical composition on the mechanical properties and on the microstructure was investigated. Totally 18 induction furnace melts were prepared by adding to a base analysis containing 0.4% carbon, 0.8% manganese and 0.1% vanadium various amounts of selected alloying elements.

Of these test melts hot forged components and round test bars were prepared. The hot forging was accomplished at 1300°C and the round bars were normalized at 1220 °C. The idea was to simulate by normalizing the hot forging and thus simplify the test procedure.

Influence of the Composition on Mechanical Properties

Mechanical properties were measured in the tensile test, and their connection with the composition was determined by using multiple linear regression analysis. As shown in Table I, the most significant alloying elements were manganese, chromium, carbon and in some cases nitrogen.
The results are presented in graphical form in Figure 1, were the tensile strength is expressed as a function of a composition sum S_1. The corresponding results for the yield strength are shown in Table II and in Figure 2. This time the most significant elements were manganese, chromium, and in some cases nitrogen. The elongation as a function of the tensile strength is shown in Figure 3, where the positive effect of a small austenite grain size is seen.

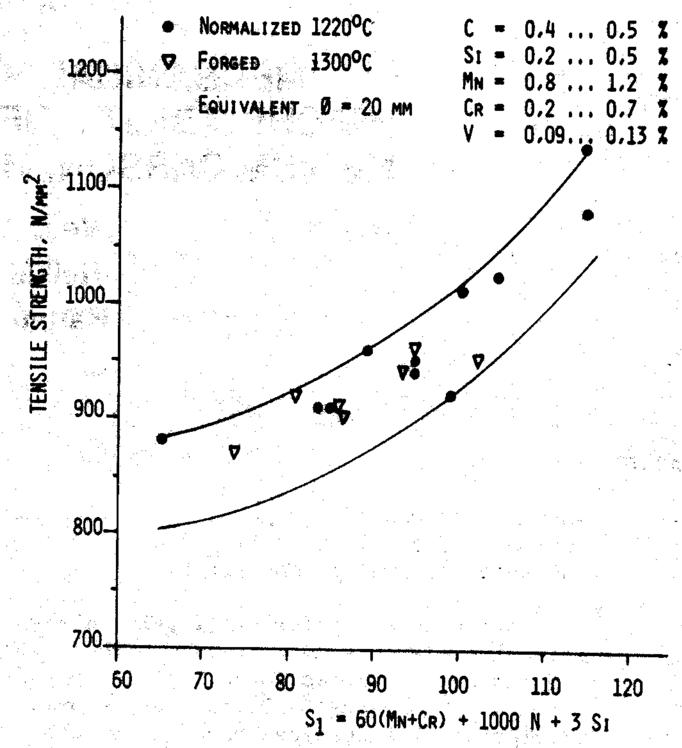

FIGURE 1 - The dependence of the tensile strength on the composition of test melts.

Condition	Regression Coefficients	Multiple Correlation	Standard Error
Normalized at 1220°C	$R_M = 503,9+347,1(Mn)^{xx} + 390,1(Cr)^{xx}$	0,87	41,8
	$R_M = 488,7+293,2(Mn)^{xx} + 312,6(Cr)^{xx} + 5708(N)$	0,91	37,9
	$R_M = 487,3+340,2(Mn)^{xx} + 347,8(Cr)^{xx} + 4320(N)^{xx}$	0,93	34,4
	$R_M = 477,8+305,4(Mn+Cr)^{xx} + 5785(N)^{x}$	0,91	35,6
	$R_M = 483,0+345,2(Mn+Cr)^{xx} + 4343(N_F)^{xx}$	0,93	32,2
	$R_M = 292,5+356,7(Mn+Cr)^{xx} + 386,4(C)^{xx} + 4791(N_F)$	0,94	30,6
Forged at 1300°C	$R_M = 634,0+253,6(Mn)^{xx} + 162,6(Cr)^{xx}$	0,96	10,4
	$R_M = 627,5+254,6(Mn)^{xx} + 156,0(Cr)^{xx} + 514,8(N)$	0,96	11,3
	$R_M = 334,8+157,3(Mn+Cr)+ 828,2(C)^{xx} + 2388(N)^{x}$	0,97	9,5

^{xx} or ^{x} = Degree of Significance

$N_F = N - Al/2$

Equivalent Round Bar Thickness 20 mm

TABLE I - Multiple regression analysis-tensile strength/composition.

Condition	Regression Coefficients	Multiple Correlation	Standard Error
Normalized at 1220°C	$R_E = 111.2+405.0(Mn)^{XX} + 453.4(Cr)^{XX}$	0.85	52.5
	$R_E = 97.91+357.6(Mn)^{XX} + 385.3(Cr)^{XX} + 5014(N)$	0.88	52.1
	$R_E = 95.12+398.3(Mn)^{XX} + 412.4(Cr)^{XX} + 4187(N_F)^{X}$	0.89	48.8
	$R_E = 55.29+428.2(Mn+Cr)^{XX} + 3176(N)$	0.92	44.5
	$R_E = 87.26+407.6(Mn+Cr)^{XX} + 4230(N_F)^{X}$	0.89	45.7
Forged at 1300°C	$R_E = 377.2+175.0(Mn)^{XX} + 56.26(Cr)^{X} + 3985(N)^{XX}$	0.96	10.1
	$R_E = 430.6+157.0(Mn)^{XX} + 60.45(Cr) + 2842(N_F)$	0.80	20.1
	$R_E = 359.5+173.8(Mn)^{XX} + 59.69(Cr)^{XX} + 4161(N)^{XX} + 51.13(Si)$	0.98	8.95

^{XX} or ^{X} = Degree of Significance

$N_F = N - Al/2$

Equivalent Round Bar Thickness 20 mm

TABLE II - Multiple regression analysis-yield strength/composition.

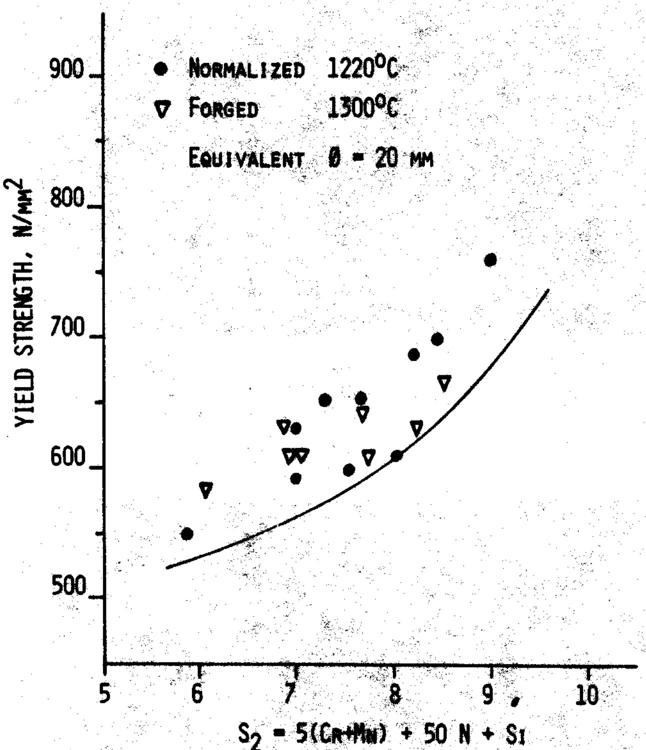

FIGURE 2 - The dependence of the yield strength on the composition of test melts.

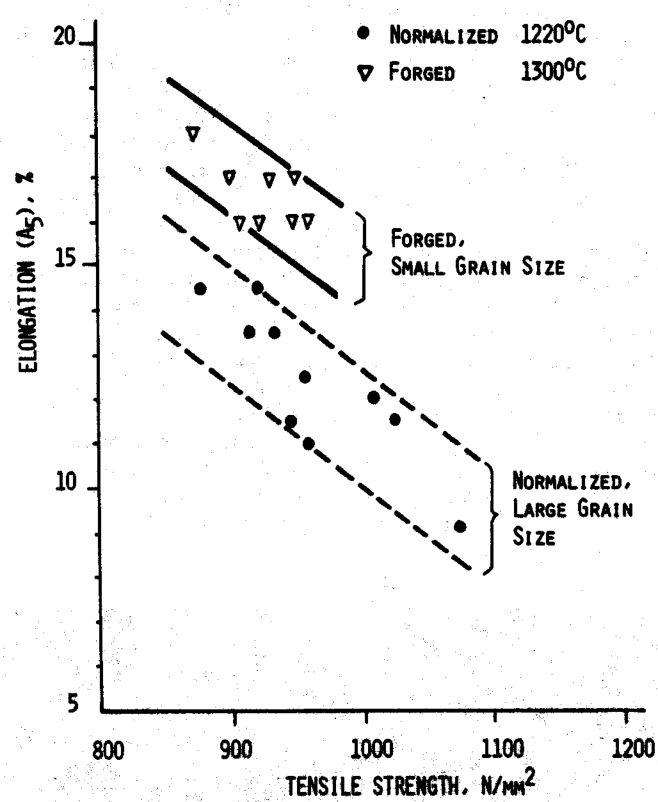

FIGURE 3 - Elongation of test melts.

Microstructures

A ferrite network revealing prior austenite grain boundaries was typical for the high strength melts alloyed with manganese or chromium + manganese, Figure 4. In SEM examination was stated that chromium + manganese alloying led to a coarser pearlite than a plain manganese alloying. An example measured from a hot forged component was 0.38 µm interlamellar distance for a 1% Mn - 0.4% Cr steel and 0.30 µm respectively for a 1.3% Mn steel.

In TEM examination dark field technigue was used due to the weak diffraction of V(C,N) precipitates. These were supposed to exhibit the Baker-Nutting orientation.

In the base composition (0.45% C - 0.8% Mn-0.1% V) V(C,N) particles were mostly randomly distributed in proeutectoid ferrite, and some of them being precipitated at dislocations, Figure 5. Large (> 20 nm), discoherent particles, which did not always exhibit the Baker-Nutting orientation, were assumed to have been precipitated in austenite subgrain boundaries after hot working. Small (~5 nm), coherent carbonitrides (Figure 5) have evidently precipitated in ferrite, although observations of coherent γ-precipitation has been reported in the literature [1]. In pearlite both randomly distributed and row precipitated carbonitrides were found embedded in ferrite. In some specimens presumably γ- nucleated V(C,N) particles were observed in various pearlite grains at the same V(C,N) reflection.

In manganese + chromium alloyed high strength test melts carbonitrides were smaller and their density higher than in the base composition, Figure 6. Regular interface row precipitation was dominant both in proeutectoid ferrite and in pearlite. However, both random and fibrous interface precipitation had occured. Thus, it can be presumed, that V(C,N) has precipitated mainly during and after γ/α transformation. An example of fibrous precipitation within a ferrite grain is shown in Figure 7. Short fibres lying almost perpendicular to the rows is seen indicating the growth during transformation parallel to the moving γ/α interface. An increase in particle size and a change in the precipitation mode from row to random were noticed in one specimen near a ferrite grain boundary, Figure 8, which is reported also in the literature [2]. Seemingly chromium and manganese have affected the precipitation kinetics thus raising the strength also through increasing the amount of precipitation strengthening.

FIGURE 4 - Ferritic-pearlitic structure of the high strength melt.

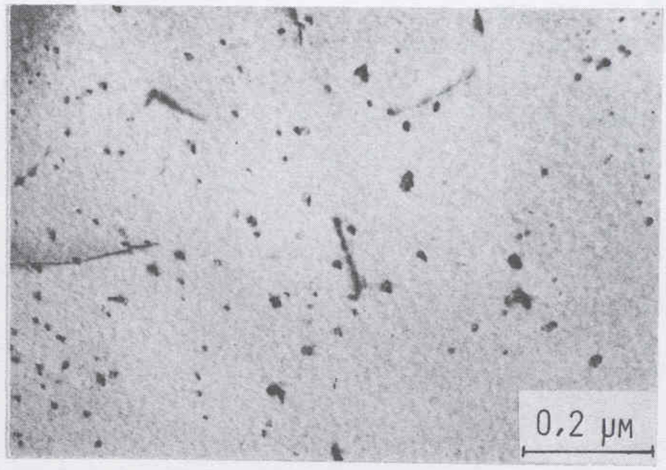

FIGURE 5 - Randomly distributed V(C,N) precipitates in the base composition.

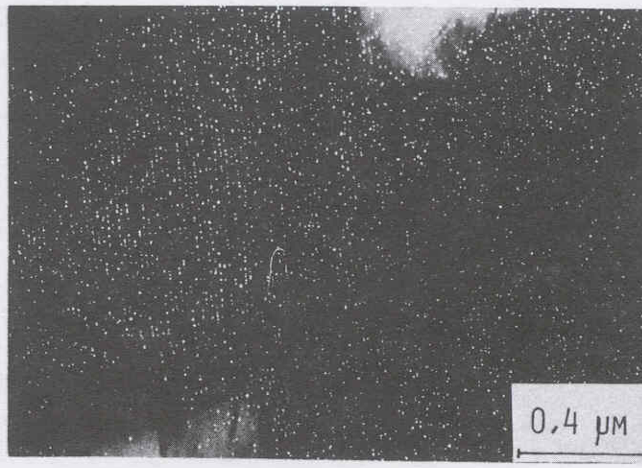

FIGURE 6 - Row precipitation in manganese + chromium alloyed high strength melt.

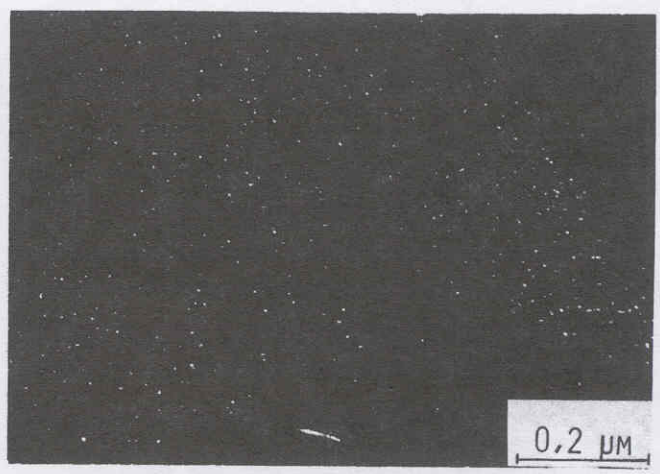

FIGURE 7 - Row and fibrous interface precipitation in high strength melt.

In a 0.1% V - 0.03% Nb alloyed, high strength test melt small, randomly precipitated particles were found at some ferrite grain boundaries Figure 9.

In certain conditions, bainite and martensite formation is possible in these steels, examples of them are shown in Figure 10 and 11. Excessive amounts of manganese(>1%)seemed to favour the formation of bainite and martensite in the microstruture.

Preliminary Specification

As the result of the laboratory scale development, the high strength grade was preliminary specified. The beneficial type of alloying would be manganese + chromium, comprising of a coarser interlamellar spacing and a smaller risk for bainite and martensite formation as plain manganese alloying. The lowered toughness and machinability could be expected due to bainite and martensite. The following nominal composition for the high strength grade was selected : carbon 0.45%, silicon 0.5%, manganese 1%, chromium about 0.5% and vanadium 0.1%. With this composition a tensile strength of roughly 900 N/mm^2 minimum and a yield strength of roughly 600 N/mm^2 minimum are obtained.

PRODUCTION SCALE DEVELOPMENT

After the first stage of the development program, toughness of the high strength grade,especially under impact loading, required closer investigation. Further, since most service fractures of machine parts are today caused by fatigue, the reliability of the high strength grade in alternating stress conditions had to be examined. When substituting for Q&T steels, the high strength grade should have at least the same machinability. Due to the sensitiveness of the fatigue resistance and machinability to the non metallic inclusions, the second stage of development had to be carried out with production scale melts.

FIGURE 8 - V(C,N) precipitates near grain boundary in high strength melt.

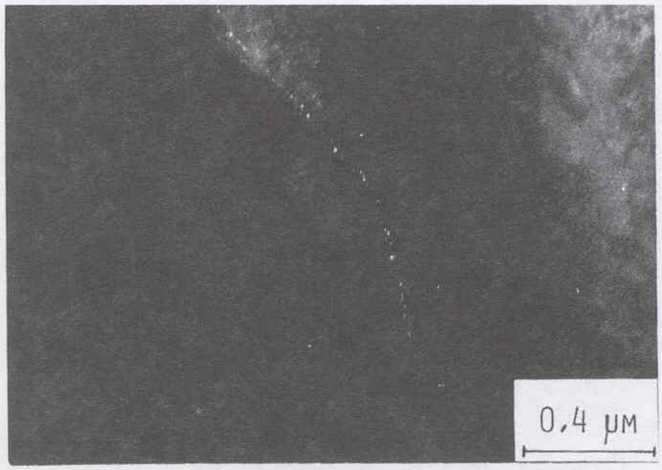

FIGURE 9 - Precipitates at grain boundary in vanadium + niobium alloyed high strength melt.

FIGURE 10 - Bainite in high strength melt.

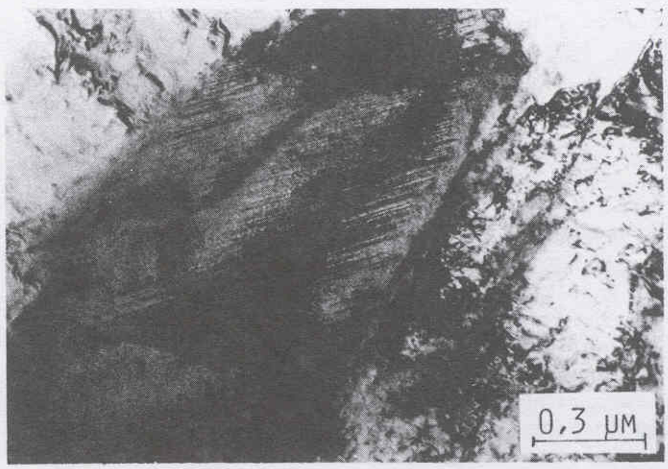

FIGURE 11 - Plate martensite in high strength melt.

Ductility and Impact Strength

The toughness of medium carbon ferritic-perlitic steel can generally be influenced without deteriorating the strength:

(i) by improving the toughness of ferritic-matrix;

(ii) by avoiding bainite and martensite; and

(iii)by decreasing the austenite grain size

(i) Ideas have been borought forward that silicon could improve the ductility and impact strength of ferritic-pearlitic matrix[3,4], probably based on the greater amount of ferrite[4].

(ii) Bainite and martensite are not desired constituents in these steels. In addition to the earlier mentioned manganese, factors favouring bainite and martensite formation are rapid cooling and large austenite grain size.

(iii) Austenite grain size can be minimized through the influence of the composition and the fabrication practice. Possible compositional elements are e.g. aluminium, niobium, and titanium. Aluminium is attractive since it does not cause the dependence of strength on the hot working temperature as niobium and titanium do. With niobium is attributed unhomogenous precipitation mode[4] (Figure 9) and possible toughness problems. Important fabrication factors influencing the grain size are the hot working temperature, especially the initial heating temperature, and the cooling rate after working. There is also the possibility of normalizing the hot worked component afterwards at a low temperature.

In Figure 12 the results are shown of the influence of high normalizing temperature (to simulate hot forging) on the mechanical properties. Temperatures between 1150 °C and 1300 °C have only a slight influence on the strength. The drop in the strength between 1150°C and 1220 °C is probably due to the increase in the grain size. The rise in the strength between 1150 °C and 1300 °C is likely caused by the decrease in the ferrite content; the ferrite network was discontinuous after normalizing at 1300 °C. An increase in the temperature decreases ductility (elongation, reduction of area) and impact strength, which is most likely attributed with a coarser grain size with increasing temperature. The two melts in Figure 12 differ in the amount of alloying (S_1) and aluminium content.

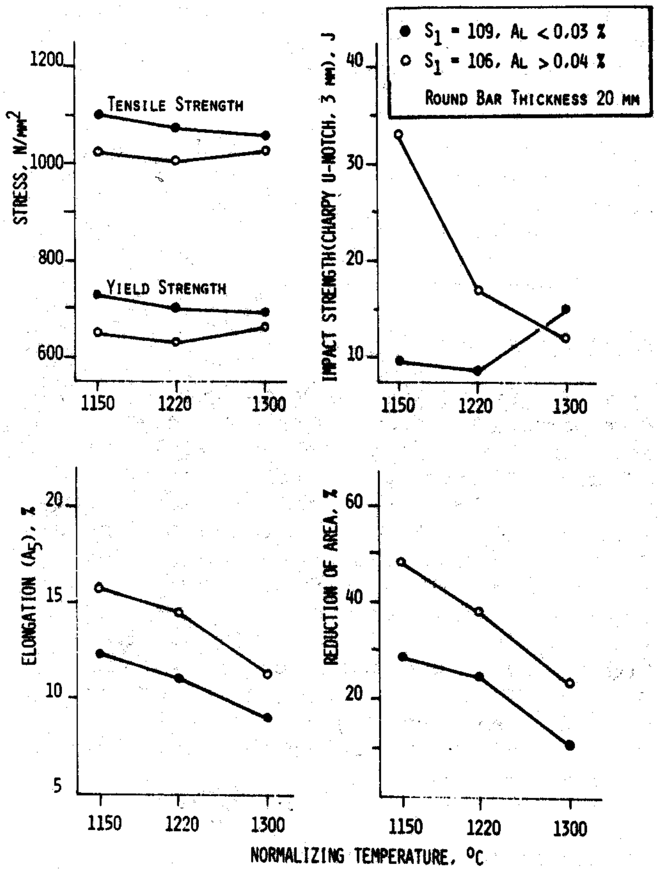

FIGURE 12 - Influence of high temperature normalizing on mechanical properties of the high strength grade.

The higher ductility and impact strength of the lower alloy melt is attributed, in addition to the lower strength level, with the higher aluminium content.

Bainite and martensite increase the tensile and yield strength and enlarges the scatter of the yield strength as shown in Figure 13. Comparatively poor yield strength values are possible at the highest bainite + martensite contents. This may be due to the lack of precipitation strengthening (Figures 10 and 11). High tensile strength values(>1100 N/mm^2) may exist even low bainite + martensite contents. This means poor machinability which is stated also in the literature[5]. Bainite and martensite lessen toughness, especially ductility, but the effect on the impact strength is not as evident.

The low temperature normalizing after hot working was simulated by a double normalizing consisting of the first normalizing at 1220 °C and the second normalizing between 950 °C and 1100 °C.

A different type of ferritic-pearlitic structure is found in the low temperature normalizing compared with usual high temperature treatments. There is a greater amount of ferrite and its distribution is not network type, but more irregular as seen in Figure 14 . Figure 15 shows that increasing (the second) normalizing temperature raises strength values but lowers ductility and the impact strength. The horizontal broken lines in Figure 14 show the values recorded after the first normalizing at 1220 °C. When the values after each normalizing are compared with each other, it can be concluded, that a low temperature normalizing at 1050 - 1100 °C leads to the same strength as normalizing at 1220 °C (i.e. hot forging) and at the same time to an improved ductility and impact strength.

The ductility and impact strength of the high strength grade as a function of average austenite grain size is shown in Figure 16. The results were obtained after different heat treatments of melts with various aluminium contents.

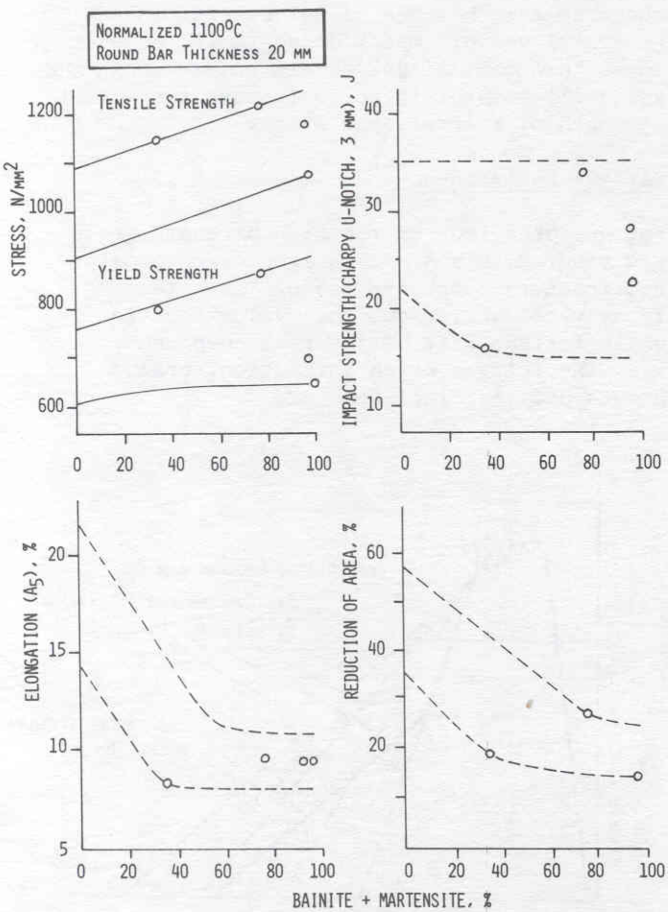

FIGURE 13 - Influence of bainite and martensite on mechanical properties.

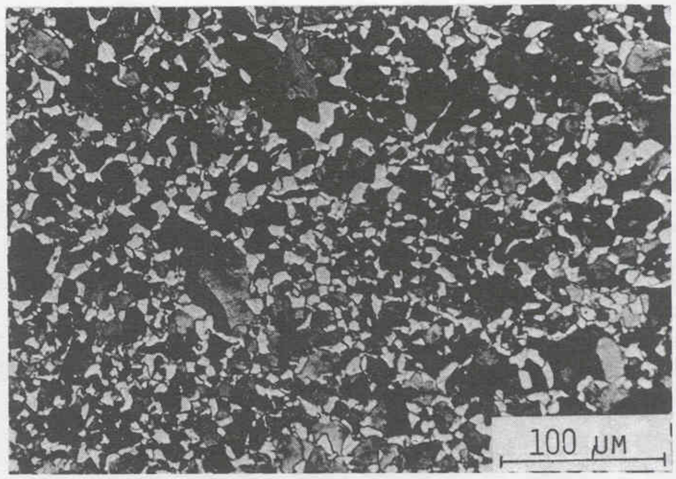

FIGURE 14 - Ferritic-pearlitic structure of the high strength grade after low temperature normalizing at 1050 °C.

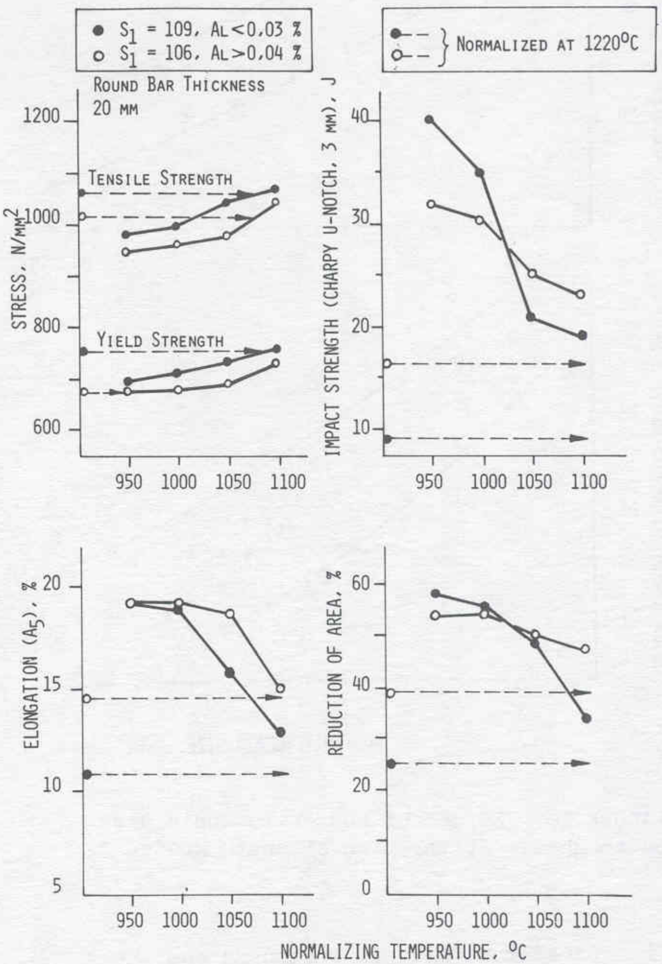

FIGURE 15 - Influence of low temperature normalizing on mechanical properties of the high strength grade.

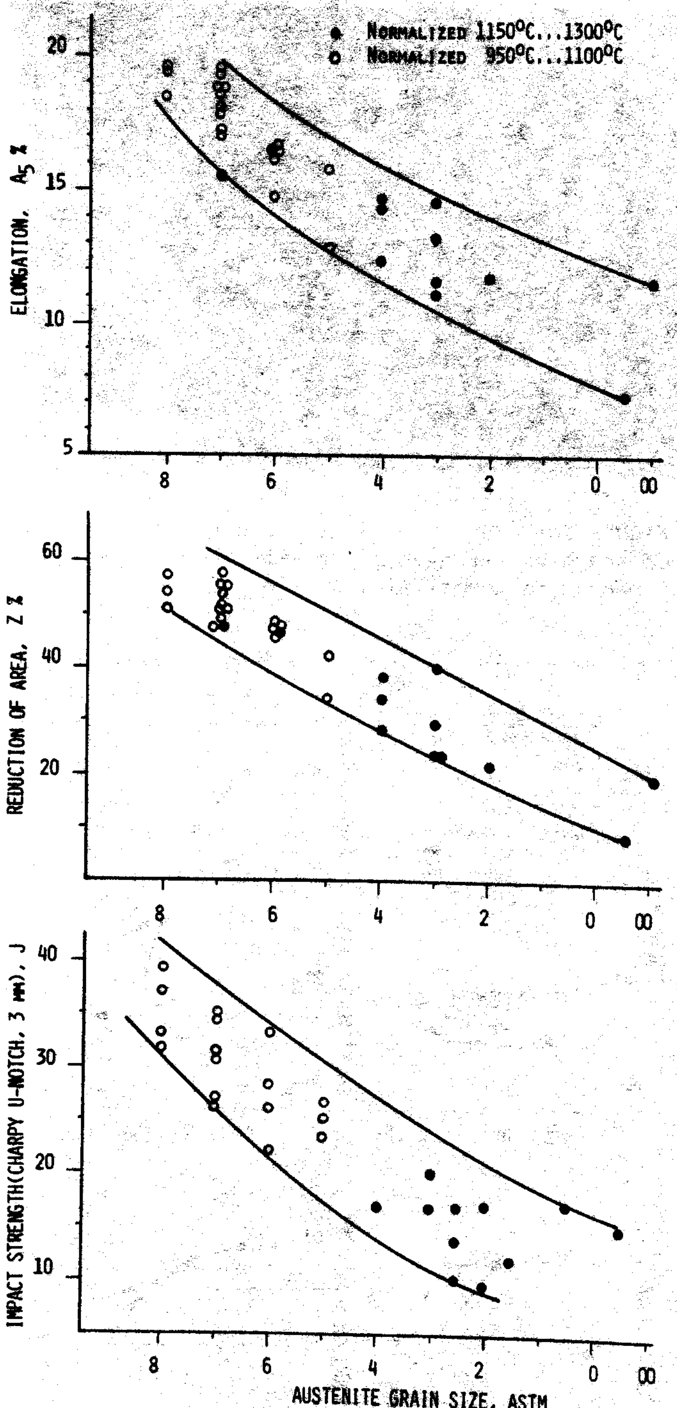

FIGURE 16 - Effect of austenite grain size on toughness of the high strength grade.

The correlation between toughness and the austenite grain size is distinct. Efforts to improve toughness should concentrate on reducing the austenite grain size.

A reasonable method to improve the toughness would be modifying the cooling after hot working. In a schematic time-temperature diagram, shown in Figure 17, different constant cooling rates are drawn corresponding to individual air cooling of different thicknesses of the component. An optimum cooling history might be as follows: at high temperatures after the finish hot deformation and above the start of pearlite reaction, i.e. above 600 - 700 °C, the cooling of thicker dimensions should be accelerated to prevent grain growth. For small dimensions it is beneficial to retard the cooling after the start of pearlite reaction to avoid residual stresses, too small interlamellar spacing or even bainite and martensite formation. An example of the effect of cooling rate on mechanical properties are presented in Figure 18. As regards toughness an optimum diameter of about 30 mm seems to exist. Below the optimum thickness, the strength increases reducing toughness (amount of ferrite and interlamellar spacing decrease, precipitation strengthening increases). Above the optimum diameter the strength remains constant, but due to a coarser austenite grain size toughness is again reduced.

Generally the static ductility values of properly treated microalloyed steel are comparable with those of Q&T steels. The general level of impact strength is, however, lower than that of Q&T steels but as shown there are still possibilities to improve the impact strength of microalloyed steel.

Fatigue Resistance

Fatigue behaviour of the high strength grade was studied in Wöhler, fatigue crack growth, and fracture toughness tests. Three test types were carried out for evaluating the whole fatigue life of the real component, i.e. the fatigue crack initiation, crack growth and the final fracture.

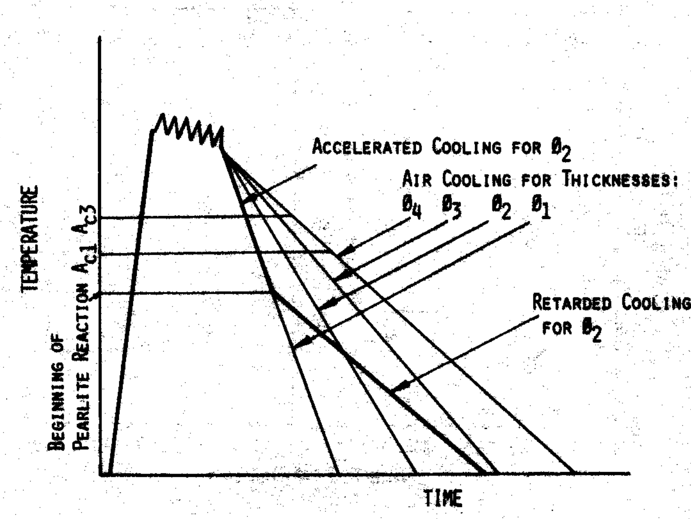

FIGURE 17 - Modified cooling after hot working (schematically).

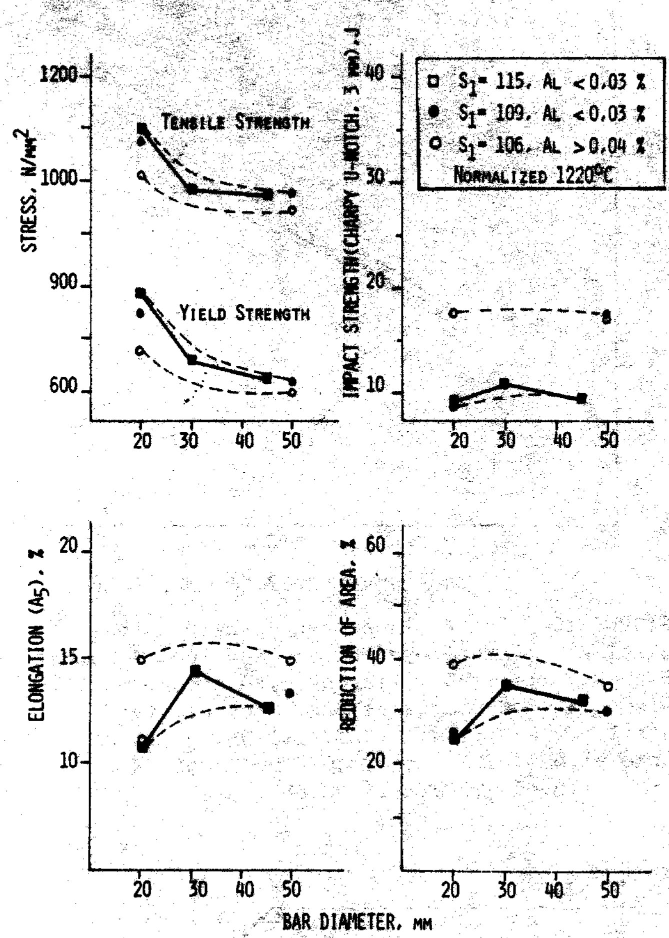

FIGURE 18 - Influence of bar thickness on mechanical properties of the high strength grade.

The crack initiation stage was tested with smooth, axially loaded specimens in reversed tension-compression. Fatigue strength at 50% fracture probability was measured using staircase method up to 10^6 cycles. According to the results, Table III, the ratio of fatigue strength to tensile strength is clearly higher (10 to 23%) for high strength microalloyed steels than for Q&T steels. The higher fatigue strength has been confirmed with full sized components, too. The more difficult crack initiation of high strength microalloyed steels was qualitatively noticed during the precracking of fracture toughness specimens. With normal low strength microalloyed steel has not been found as high fatigue strength to tensile strength ratios. Uneasy fatigue crack initiation in the high strength grade is likely attributed with the characteristic precipitation structure of the high strength grade: the fine particle size and short interprecipitate distance (Figure 6 and 7).

Fatigue crack growth test were carried out with precracked CT type specimens. Crack growth was followed microscopically on the polished surfaces. The results in the form of Paris lines are presented in Figure 19.

The difference in crack growth rate between microalloyed steels and the reference Q&T steel is negligible being of the order of the scatter in the measured values.

STEEL	TREATMENT	FATIGUE STRENGTH N/MM^2	TENSILE STRENGTH N/MM^2	ELONGATION (A_5) %	FATIGUE STRENGTH / TENSILE STRENGTH
MICRO-ALLOYED S_1 = 115 AL < 0.03 %	HOT ROLLED	492	991	19	0.50
	NORMALIZED 1220°C	503	1099	11	0.46
	NORMALIZED (1220+) 1050°C	495	1016	17	0.49
MICRO-ALLOYED S_1 = 106 AL > 0.04 %	HOT ROLLED	492	937	18	0.52
CRMO-ALLOYED Q&T STEEL	Q&T	398	939	16	0.42

TABLE III - The results from the high cycle fatigue test (crack initiation stage).

The plain strain fracture toughness of the high strength grade was determined with CT specimens using linear elastic method. In many cases there happened a pop-in, i.e. a small unstable crack growth. In the case of pop-in, calculated fracture toughness values were formally invalid, for the reason shown schmatically in Figure 20. However, all the measured values are presented in Table IV.

Since valid values for Q&T steel were not obtained with the used specimen size, J integral tests for these steels were carried out using the compliance method. Fracture toughness values (K_{Ic}) calculated from J integral are shown in Table V. According to the results (Tables IV and V) the fracture toughness of the high strength microalloyed steel is poorer and the critical crack size smaller than those of CrMo alloyed Q&T steel.

Because of its smaller critical crack size, a shorter fatigue life could be expected for microalloyed steel.
However, in many stress states, e.g. in fluctuating tension, the the crack growth rate accelerates rapidly with increasing crack size, schematically presented in Figure 21. Thus the fatigue crack growth life of the microalloyed steel might be only slightly smaller than that of Q&T steel. Due to the longer crack initiation stage (Table III) the whole fatigue life of microalloyed steel may even be longer than that of Q&T steel.

The higher fatigue strength (Table III) means that for many components the tensile strength of microalloyed steel may be lower than the tensile strength of Q&T steel to be substituted. This point of view should be taken into consideration when other properties, strongly influenced by hardness or tensile strength (such as machinability and impact strength), are compared between microalloyed and Q&T steels.

Machinability

The machinability of the high strength microalloyed steel has been tested in different operations.

Millability was investigated in Volvo' (SFA) standard end milling test with a H.S.S. cutter. The results in Figure 22 show, that at lower hardness levels the millability of microalloyed steel is better than that of CrMo alloyed Q&T steel. At higher hardness levels corresponding to the high strength grade the difference gradually vanishes. However, the millability of high strength grade is at least the same as that of CrMo Q&T steels.

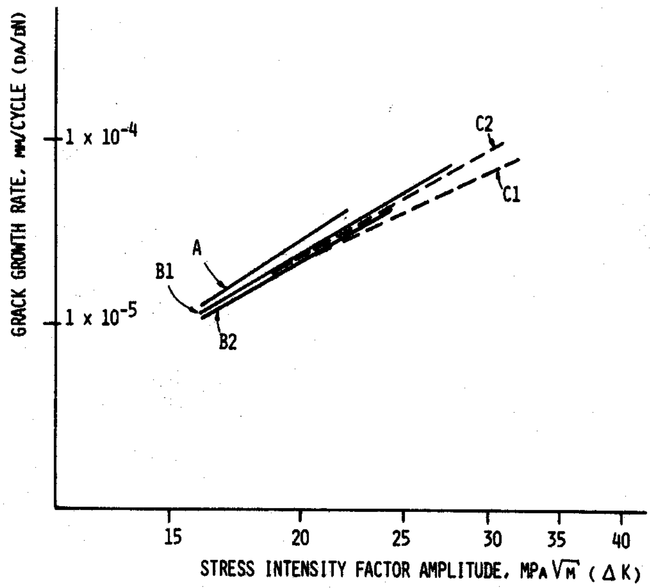

FIGURE 19 - The fatigue crack growth rates. Steel A = microalloyed steel, S_1 = 106 Al>0.04%, 305 HB. Steel B1 = microalloyed steel, S_1 =115 Al<0.03%, 316 HB. Steel B2 = microalloyed steel, S_1 = 115, Al<0.03%, 298 HB. Steel C1 = CrMo alloyed Q&T steel, 262 HB. Steel C2 = CrMo alloyed Q&T steel, 320 HB.

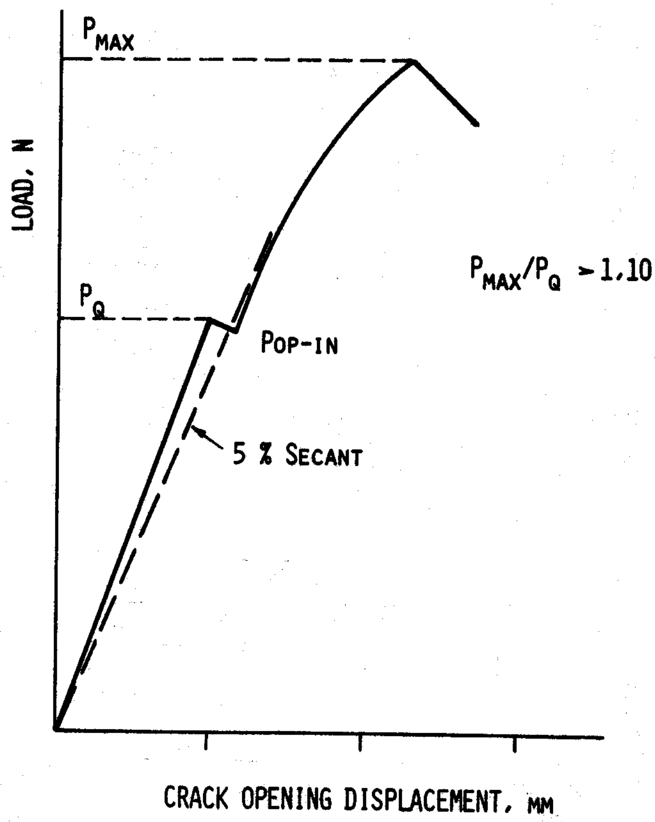

FIGURE 20 - A schematic load - crack opening curve with pop-in. According to ASTM E 399-74 P_{max}/P_Q being over 1.10, the measured fracture toughness value is invalid.

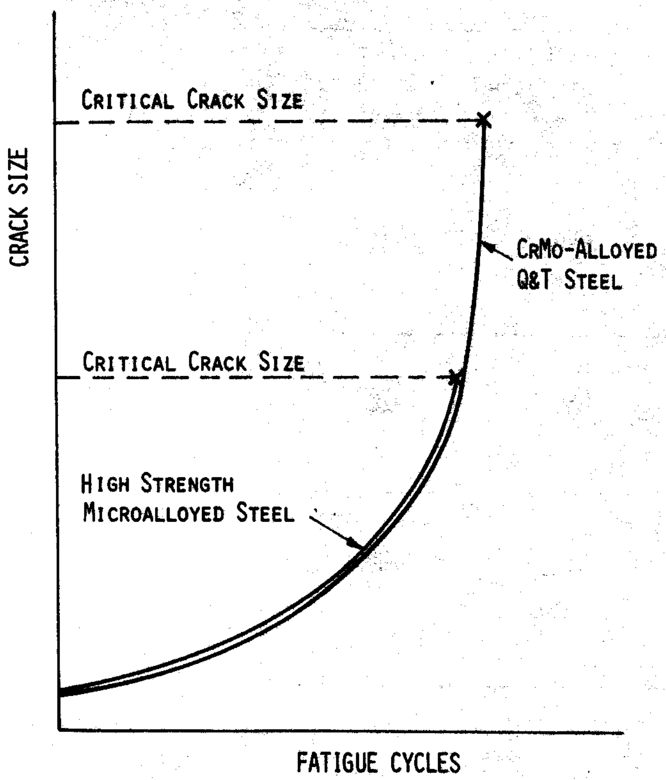

FIGURE 21 - A schematic curve of the fatigue crack growth and final fracture of the high strength microalloyed steel and CrMo alloyed Q&T steel.

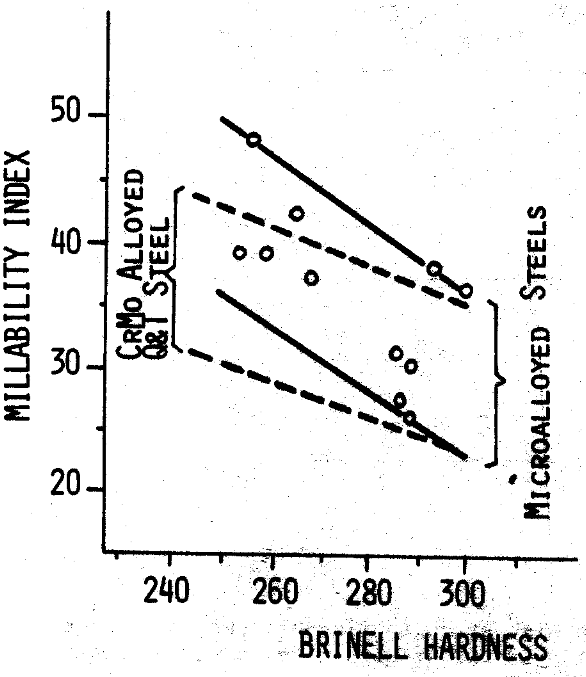

FIGURE 22 - Millability of microalloyed and Q&T steels.

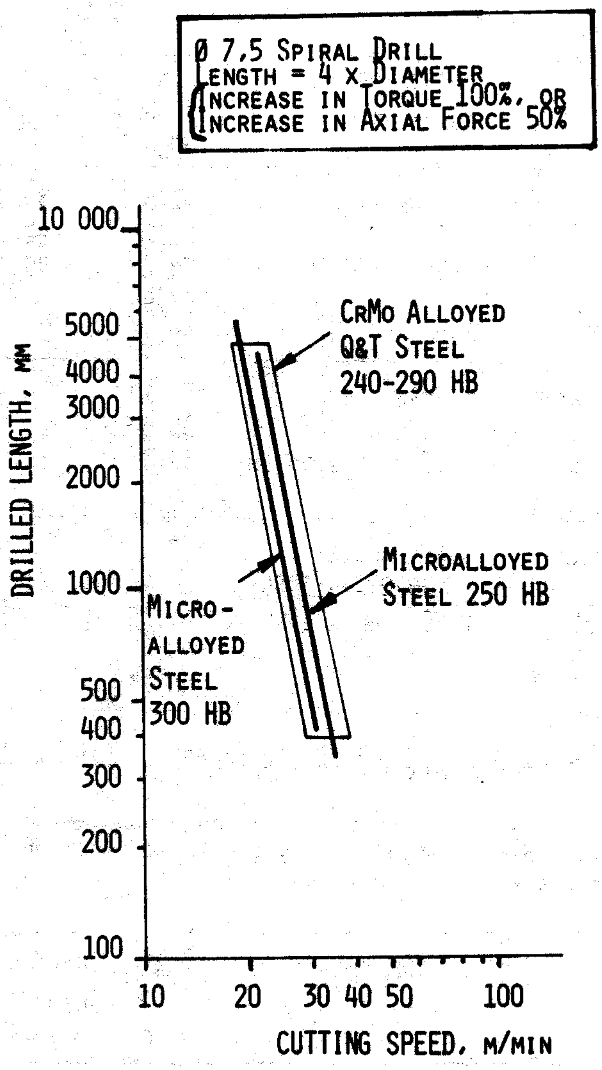

FIGURE 23 - Drillability of microalloyed and Q&T steels.

For the smallest holes (d < 4 mm) a special type of deep hole drills with a sufficiently wide chip disposal channel is recommended The microalloyed steel is also sensitive to the cutting data and quality of cutting fluid in deep hole drilling.

Single point turning test with cemented carbide tools were carried out according to ISO 3865. In finish turning the tool wear was smaller for the microalloyed steels, especially at higher hardness levels, than for the CrMo alloyed Q&T steels, Figure 24. The surface finish of the high strength grade has also been proved to be excellent. In rough turning no essential difference was found between a normal high strength microalloyed steel and CrMo alloyed Q&T steel at the same hardness level, Figure 25. Calcium treatment, when leading to round calcium aluminate oxide inclusions enveloped by sulphides, has proved to have an overall beneficial effect on the machinability both with H.S.S. and carbide tools [,8].

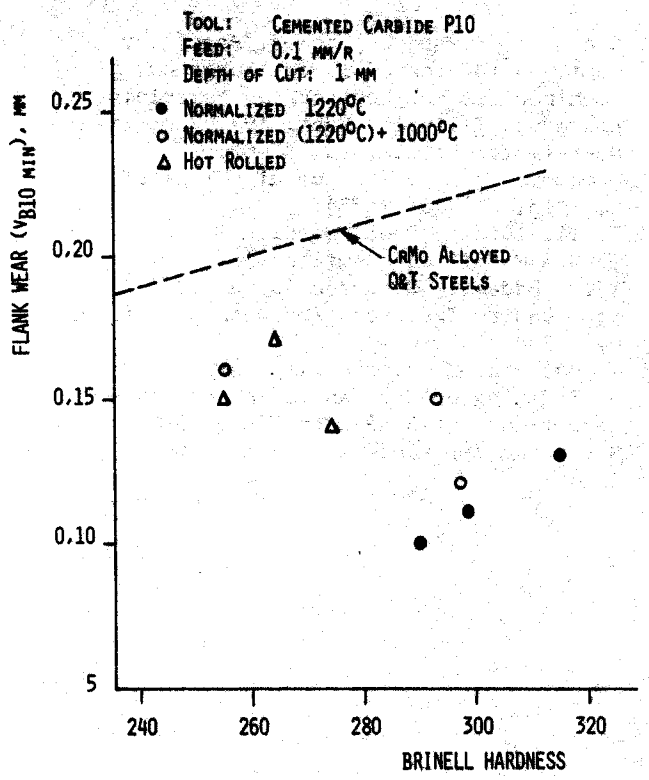

FIGURE 24 - Wear of carbide tool in finish turning of microalloyed and Q&T steels.

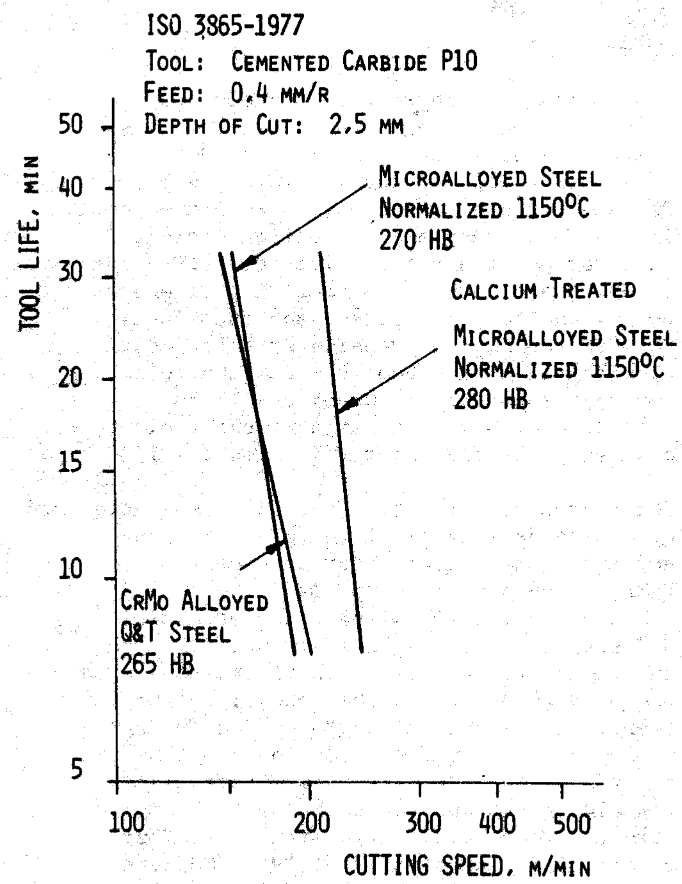

FIGURE 25 - Taylor lines in rough turning for Q&T steel, high strength microalloyed steel, and for calcium treated (M-treated) microalloyed steel.

Example of the effect of calcium treatment (M-treatment) in rough turning of high strength microalloyed steel is also seen in Figure 25.

Specifying the High Strength Grade

A high strength microalloyed steel, called IVA 1000, has been developed. Its nominal composition is the same as was mentioned:

0.45% C, 0.5% Si, 1% Mn, ~0.5% Cr, 0.1% V

The mechanical properties are normally specified in detail by the supplier in collaboration with the user. An example of a specification is presented in Table VI. Because the final properties depend on the dimensions of the component and on the forging practice, the composition-mechanical properties combination should be reconsidered for every specific component. An improved toughness is possible by modifying composition and forging practice or by low temperature normalizing. Calcium treatment (M-treatment) is recommended for microalloyed steel for an overall improved machinability.

The use of the high strength grade is divided into two lines: substitution of Q&T steels at the same strength level and substitution of lower strength steels for weight saving, e.g. ordinary microalloyed steel.

YIELD STRENGTH N/mm^2 MIN	TENSILE STRENGTH N/mm^2	ELONGATION (A_5) % MIN	REDUCTION OF AREA % MIN
625	930-1080	12	20

TABLE VI - Example of mechanical property specification of IVA 1000.

The first mentioned line means that no essential reduction in other properties than impact strength are expected.

In the latter line the machinability, especially with H.S.S. tool, is likely reduced. At the moment IVA 1000 is being tested in different automotive components, e.g. in connecting rods, spindles, crankshafts and driveshafts.

SUMMARY

The purpose of the work was to develop a high strength microalloyed steel with a minimum tensile strength of about 900 N/mm^2 being able to substitute for alloy Q&T steels. A number of melts were made both in the laboratory and production scale and their properties tested. The most effective alloying elements to increase the strength were found to be manganese and chromium. The desired minimum tensile strength is obtained after air cooling with the following type of composition: 0.45% C - 1% Mn - 0.5% Cr - 0.1% V. This high strength micro-alloyed grade developed is called IVA 1000.

The microstructure of IVA 1000 is characterized by a high density of V(C,N) precipitates and their small size. Most service and fabrication properties of IVA 1000 are comparable with Q&T steels. The essential differences were the lower impact strength and higher fatigue strength of the high strength grade. However, there are possibilies to improve the impact strength by modifying the composition and hot working practice or by applying low temperature normalizing. The higher fatigue strength may be attributed with the fine precipitate structure. Most machinability properties of IVA 1000 were equal or better than those of Q&T steels, although a different behaviour in deep hole drilling was observed. Calcium treatment (M-treatment) is recommended for an overall improved machinability. The high strength grade is used in automotive component not only to substitute for Q&T steels but also for ordinary microalloyed steel when weight saving is desired.

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