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Development of a New Direct Quenched Abrasion Resistant Steel

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Abstract A new way of producing abrasion resistant steel with a hardness substantially greater than 500 HV via direct quenching has been investigated. To avoid the ductility problems associated with fully martensitic microstructures, we investigated the possibility of achieving the desired hardness by incorporating a small amount of soft ferrite in a martensite matrix. The approach used was to study the potential suitability of several candidate compositions with varying levels of aluminium and manganese using physical simulation and laboratory rolling combined with metallography and abrasive wear testing. Physical simulation included the determination of CCT diagrams for strained and unstrained austenite on a Gleeble thermo-mechanical simulator. The CCT diagrams quantified 1) the increase in ferrite formation temperatures as a function of aluminium content, 2) the level of alloying required to obtain fully martensitic microstructures at appropriate cooling rates, and 3) the influence of prior austenite straining on the ferrite and bainite transformation kinetics. On the basis of the simulation results, laboratory rolling trials were made. They showed that a hardness even above 600 HV could be obtained with the desired martensitic-ferritic microstructure by designing a suitable composition and selecting appropriate finish rolling and quench start temperatures. Abrasive wear testing was used to compare the wear resistance of the laboratory rolled material with that of commercial direct quenched abrasion resistant grades.

Keywords Steel, Abrasive Wear, Multiphase Microstructure, Physical Simulation

1. Introduction

Abrasive wear is a severe problem in the mining, earth moving and transportation industry, and huge savings could be made if suitably priced and easily fabricated materials with improved resistance to abrasive wear were available[1]. Steel is the most commonly used abrasion resistant material, and since the life of abrasion resistant steel is largely dependent on hardness, there is a desire to find ways of making and using steels with as high hardness as possible. Commercially available abrasion resistant steel grades are often fully martensitic with a hardness of 400, 450 or 500 HB (420 to 527 HV). During recent years, there has been an increasing interest in the use of abrasion resistant steel produced via direct quenching where austenite is quenched to martensite immediately following hot rolling without the need for intermediate cooling and reaustenitization.

With hardnesses in the range 400 – 500 HB, relatively good combinations of abrasion resistance, toughness, cutability, bendability, weldability and resistance to hydrogen induced cracking can be obtained. However, as the hardness increases, these technological properties deteriorate and the production and use of the steel become ever more demanding. Consequently, it is difficult to profitably produce and utilize martensitic steels with a hardness higher than 500 HB (527 HV, 51 HRC).

It has been postulated that the low ductility and toughness associated with overall hardness levels above 500 HB might be improved by changing the microstructure from 100% martensite to a mixture of ferrite and martensite thereby enabling more novel abrasion resistant solutions to be achieved[2,3]. The objective of the work described in this paper was to explore the practicality of such an approach when using direct quenching. This has been done by studying the potential suitability of candidate compositions with varying levels of aluminium and manganese using physical simulation, laboratory hot rolling and direct quenching together with metallography and abrasive wear testing. Physical simulation included the determination of CCT diagrams for strained and unstrained austenite on a Gleeble thermo-mechanical simulator for designing composition and laboratory hot rolling schedules.

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Table 1. Compositions in wt.% of test materials, balance Fe (S≤ 50 ppm and P≤ 60 ppm)

Cast	C	Si	Mn	Al	V	Ti	Cr	N
0.4C-0.65Mn-0.08Al-V-Cr	0.38	0.18	0.65	0.08	0.20	0.019	0.37	35 ppm
0.4C-1.49Mn-0.08Al-V	0.39	0.22	1.49	0.08	0.21	0.017	-	38 ppm
0.4C-1.15Mn-1.23Al-V	0.39	0.30	1.15	1.23	0.20	0.021	-	35 ppm
0.4C-1.50Mn-2.10Al-V	0.43	0.26	1.50	2.10	0.20	0.019	-	38 ppm
0.5C-0.77Mn-1.90Al-V-Cr	0.47	0.56	0.77	1.90	0.192	0.020	0.24	40 ppm

2. Experimental

2.1. Materials

The chemical composition of the five steels used are presented in Table 1. In order to promote ferrite formation, high levels of aluminium, silicon and vanadium were included. While V(CN) precipitates are known to be effective nucleation sites for ferrite[4,5,6], aluminium increases both the A_3 and A_1[7,8,9] and allows ferrite to form at higher temperatures and lower cooling rates[8]. Aluminium also supresses the formation of cementite[10]. Titanium was added to all compositions to form fine TiN precipitates to suppress austenite grain growth and possibly provide nucleation sites for ferrite[11,12]. Different manganese levels were included due to its effect on ferrite formation and the hardenability of the austenite. Mn slows down the decomposition of austenite and lowers the M_s temperature. MnS inclusions also can act as nucleation sites for proeutectoid ferrite[4,5,6], but sulphur levels were kept low due to the detrimental effect of MnS inclusions on impact toughness. In some cases, chromium was also added to adjust hardenability.

The experimental steels were cast as 60 kg ingots (450 x 300 x 55 mm), which were cut into (225 x 75 x 55 mm) pieces for hot rolling experiments. The starting material for the Gleeble simulations was made by hot rolling to a thickness of 11 mm and subsequent air cooling.

2.2. Dilatometry and CCT Diagrams

CCT diagrams (*Continuous cooling transformation*) were constructed for all test material compositions using dilatometric data obtained with a Gleeble 1500 thermo-mechanical simulator starting with both recrystallized and deformed unrecrystallized austenite. Test samples were 6 mm diameter cylinders machined with their axis transverse to the rolling direction.

In the Gleeble, the specimens were heated to 1100°C at 20°C/s, held at temperature for 4 min and then cooled at 2°C/s. When no deformation was applied, cooling was stopped at 1000°C where the specimens were held for 2 min. In the case of deformed austenite, cooling was stopped at 850°C for10 s for compressive deformation.

Deformation was applied in three passes each with a strain of 0.2 and a strain rate of 1 s^{-1}. The interval between each compression was 25 s and the specimens were kept at 850°C for 10 s after the third compression. Both undeformed and deformed specimens were finally cooled to room temperature at different cooling rates ranging from 1.5°C/s to 80°C/s. The diameter of the cylinders during cooling was measured using a strain gauge. For metallographic examination and hardness testing, the specimens were cut in half along the compression axis. Microstructural characterization and hardness measurements were made as explained below.

2.3. Laboratory Rolling Experiments

The aim of the laboratory rolling experiments was to find rolling process parameters suitable for obtaining a dual-phase microstructure with ferrite and martensite. On the basis of the CCT diagrams, 0.4C-1.50Mn-2.10Al-V and 0.5C-0.77Mn-1.90Al-V-Cr where chosen as the suitable materials for the rolling experiments.

All the ingot pieces were annealed at 1200°C for 2 h before hot rolling to a final thickness of 12 mm which means a total reduction 78%. Plates were rolled with all hot rolling passes completed above 1000°C, except the last one. Like other passes, the final pass too had a reduction of 0.2 strain. The temperature of the final pass (*FRT*) was 1000°C or lower (950°C, 900°C or 850°C) to complete rolling in the no-recrystallization regime. The rolled plates were finally quenched in water (*WQ*). The starting temperature of quenching was about 20-30°C lower than the FRT and the mean cooling rate between 800°C and 300°C was estimated to be about 75°C/s. For presentational purposes, the FRT is used to identify the different test plates.

2.4. Material Characterization

Microstructural characterization was made using a light optical microscope and a field emission scanning electron microscope (*FESEM*) Zeiss Ultra Plus on samples that were polished and etched with 5% Nital. Ferrite fractions were estimated from the etched samples using the point counting method with a total of 663 points from three micrographs per test material.

Hardnesses of the mounted and polished samples were measured with a Struers Duramin A-300 hardness tester. A load of 10 kg was used for rolled and quenched material and 5 kg was used for CCT specimens. The values presented are means of 10 measurements

2.5. Wear Testing

Crushing pin-on-disk testing was employed to study the resistance against abrasive wear of the most interesting test materials. Tests were carried out at the Tampere Wear Center of the Technical University of Tampere. The pin-on-disk test principle is described in the ASTM G 99 standard. The only differences were the use of a free abrasive between the pin and the disk and the presence of confining walls that kept the abrasive on the disk. The disk is 160 mm in diameter and the confining walls have an inner diameter of 150 mm. The pin, which in this case is made of the material tested, is 36 mm in diameter. The flat end area of 1000 mm^2 is pressed close to the disk with a pneumatic cylinder. The pin and the disk are set not to come into contact with each other, leaving a gap of ca. 1 mm between the pin and the disk. In the test, the pin is repeatedly pressed against a gravel bed for 5 s, followed by a 2.5 s hold in the lifted position. This procedure allows recycling of the abrasive between the pin and the disk[13]. The pin is weighed before and after the test. The mass loss of the pin is used as an indication of resistance against wear. In these tests, 500 g of uneven sized (2-10 mm) granite was used as the abrasive, the test time was 30 min and the pin pressure was 90 kPa. The disk used was varied so that the hardness ratio between the pin and the disk was kept in the range (1.00-1.16) i.e. almost constant. The mass losses of the test materials were compared to those of the commercial wear-resistant steels Raex 400 and Raex 450 provided by Ruukki Metals.

3. Results and Discussion

3.1. CCT Diagrams

In general, prior deformation of austenite accelerated the formation of ferrite so that some ferrite formed even with high cooling rates when deformation was applied leading to lower levels of hardness at a given cooling rate.

The effect of aluminium on the formation of ferrite becomes apparent when comparing the compositions 0.4C-1.49Mn-0.08Al-V and 0.4C-1.50Mn-2.10Al-V that only differ in their aluminium content (Figure 1 and Figure 2, respectively). The ferrite transformation of compositions with 2.1% Al started at temperatures just below 800°C at cooling rates from 1.5°C/s to 3°C/s (Figure 2) whereas with 0.08% Al (Figure 1) ferrite transformation started slightly above 700°C at the same cooling rates when austenite was not strained.

The effect of composition to formation of ferrite can also be seen from the hardnesses obtained with high cooling rates. For compositions 0.4C-1.49Mn-0.08Al-V and 0.4C-0.65Mn-0.08Al-V-Cr (Figure 1 and Figure 3, respectively), the hardness obtained at 80°C/s cooling rate is the same for both specimens tested with and without deformation of austenite. For other compositions, the hardness is lower when austenite is deformed, as discussed earlier.

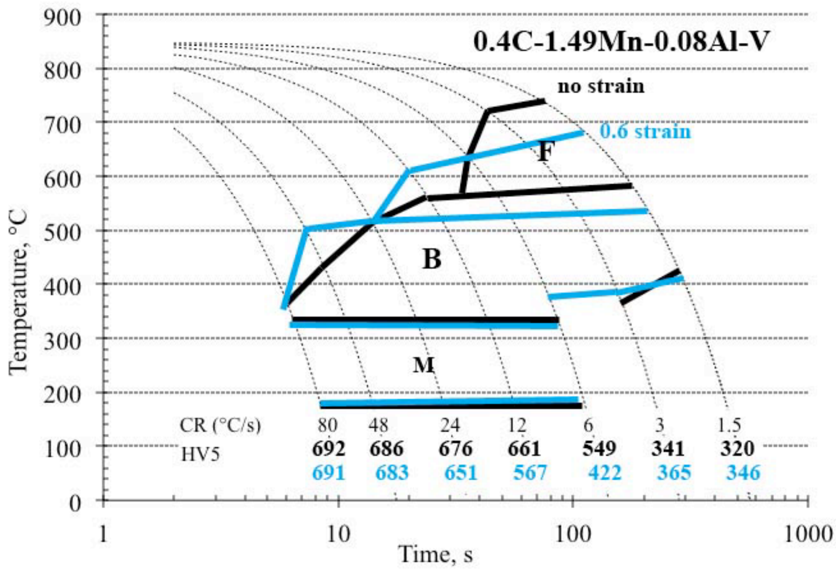

Figure 1. CCT diagram; 0.4C-1.49Mn-0.08Al-V

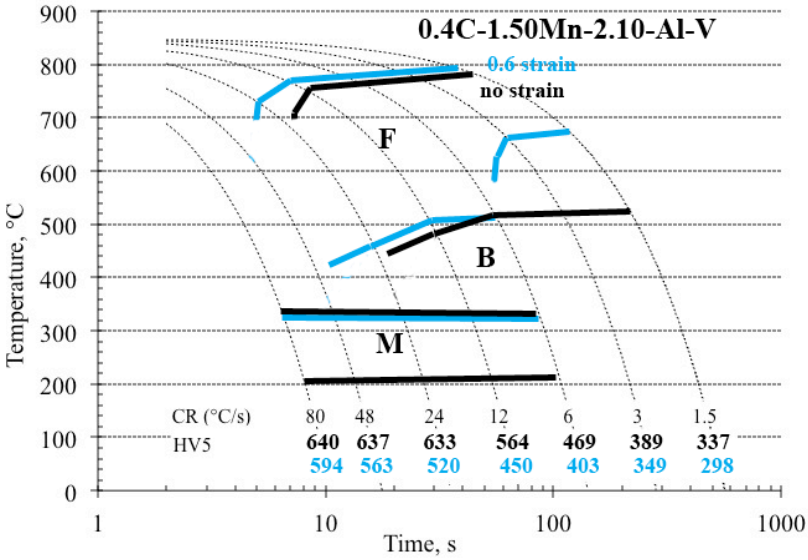

Figure 2. CCT diagram; 0.4C-1.50Mn-2.10Al-V

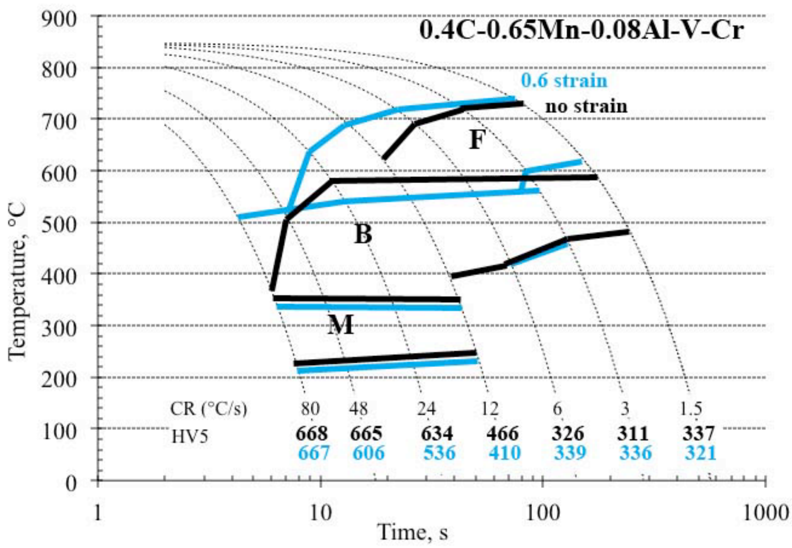

Figure 3. CCT diagram; 0.4C-0.65Mn-0.08Al-V-Cr

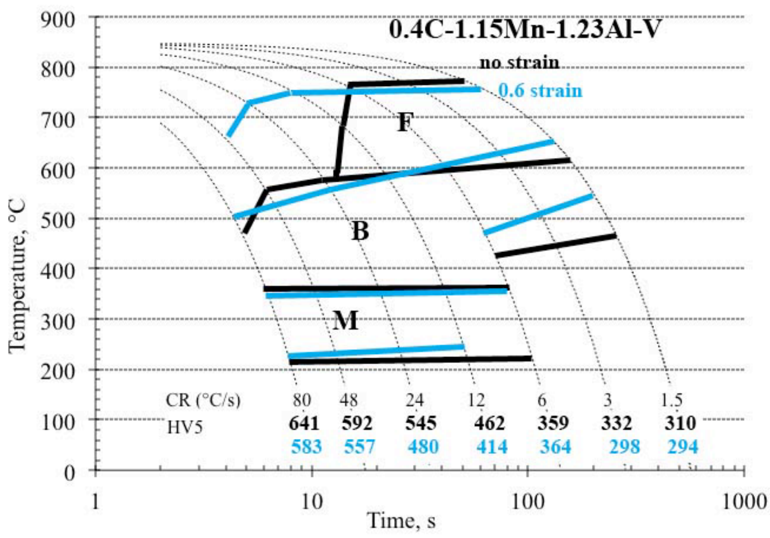

Figure 4. CCT diagram; 0.4C-1.15Mn-1.23Al-V

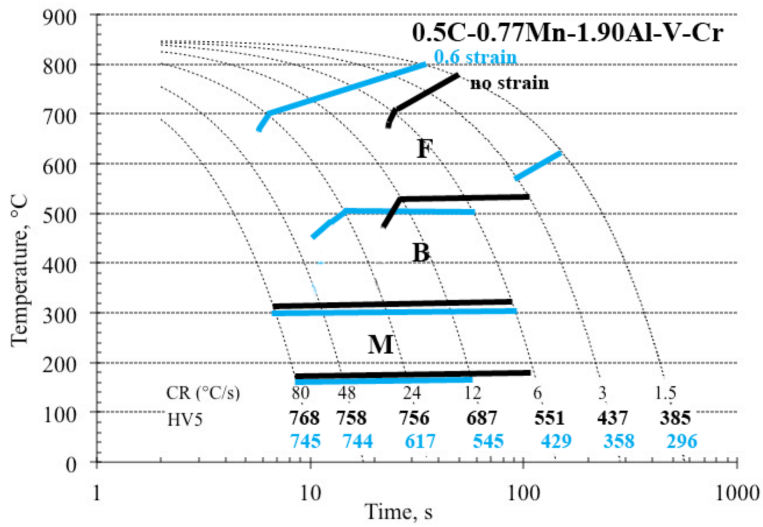

Figure 5. CCT diagram; 0.5C-0.77Mn-1.90Al-V-Cr

The effect of manganese on the phase transformation kinetics is evident when comparing the CCT diagrams of 0.4C-0.65Mn-0.08Al-V-Cr and 0.4C-1.49Mn-0.08Al-V (Figure 3 and Figure 1, respectively). Ferrite formation starts at lower temperatures in the composition with 1.49% Mn than in the one with 0.65% Mn, even though it also has 0.37% Cr, which is known to have a similar effect on hardenability as Mn. The effect is also clear in the critical cooling rates for ferrite formation: higher Mn content promotes martensitic (or bainitic) microstructure to form at lower cooling rates.

Compositions 0.4C-1.50Mn-2.10Al-V and 0.5C-0.77Mn-1.90Al-V-Cr were chosen as test materials for the rolling experiments, due to their relatively high ferrite formation temperatures combined with high hardness levels (Figure 2 and Figure 5, respectively).

Hardness of over 600 HV was obtained with both these compositions. With the carbon content of 0.5%, a martensitic microstructure containing a small amount of ferrite and a hardness of over 750 HV was attained from deformed austenite.

3.2. Laboratory Rolled Plates

Measured hardness values and corresponding proportions of ferrite in the rolled and direct quenched plates are presented in Figure 6. Plates with hardness from 483 to 734 HV were obtained. Highest hardness values were obtained with 0.5C-0.77Mn-1.90Al-V-Cr, as expected. The plates had low amounts of ferrite being almost totally martensitic. The measured hardnesses were close to the theoretical hardness of martensite, i.e. 744 HV (Table 2). In the Gleeble simulations, a marginally higher hardness of 768 HV was obtained at the cooling rate of 80°C/s (Figure 5), though the difference could, of course, simply be due to scatter.

With 0.4C-1.50Mn-2.10Al-V, relatively higher amounts of ferrite were obtained with ferrite content increasing with decreasing FRT. The highest hardness for 0.4C-1.50Mn-2.10Al-V (630 HV) was obtained with a FRT of 1000°C. Still it was significantly softer than the theoretical hardness of martensite calculated for that composition (Table 2). However, the hardness was close to the hardness obtained at 80°C/s from recrystallized austenite in the CCT tests (Figure 2). The lower hardness than the theoretical maximum is caused by a small amount of bainite in the microstructures of both the rolled plate (Figure 7c) and the CCT specimens.

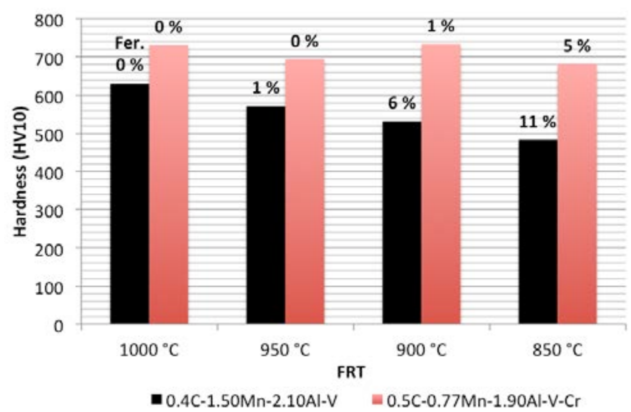

Figure 6. Hardness of the rolled and quenched plates with the amount of ferrite as a percentage

Table 2. Carbon content and theoretical hardness of fully martensitic microstructure

	C %	100%M HV
0.4C-1.50Mn-2.10Al-V	0.43	705
0.5C-0.77Mn-1.90Al-V-Cr	0.47	744

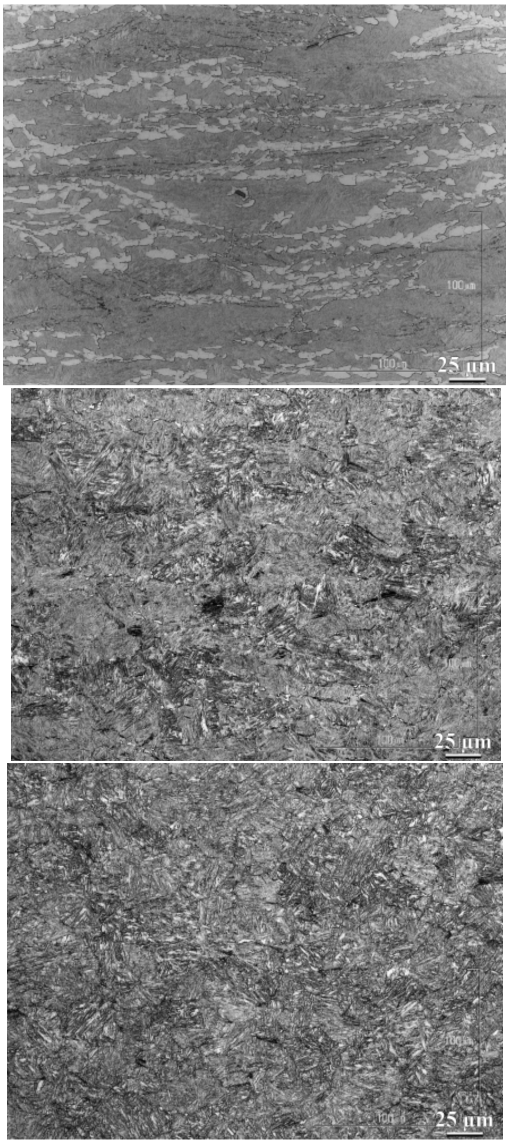

Figure 7. 0.4C-1.50Mn-2.10Al-V; a) FRT 850°C, b) FRT 900°C, c) FRT 1000°C

The effect of composition and rolling schedule on the microstructure can be seen from the micrographs presented in Figure 7 and Figure 8. From the deformed microstructure in Figure 7a, it is clear that the FRT of 850°C is below the recrystallization stop temperature (RST) for 0.4C-1.50Mn-2.10Al-V. The structure clearly shows grain boundary ferrite (about 11%). With an increase in FRT to 900°C, the microstructure reveals the formation of some bainite as well as traces of ferrite (<6%) and mild pancaking (Figure 7b). Likewise, the structure of the 0.5C-0.77Mn-1.90Al-V-Cr steel showed the formation of some ferrite (5%) at grain boundaries for specimens finish rolled at 850°C (Figure 8a). Evidence of mild pancaking in the specimen finish rolled at 900°C (Figure 8b) suggests a higher RST, at least above 900°C, for this steel. Traces of ferrite (about 1%) seen in this sample suggests that strain induced transformation has happened before cooling.

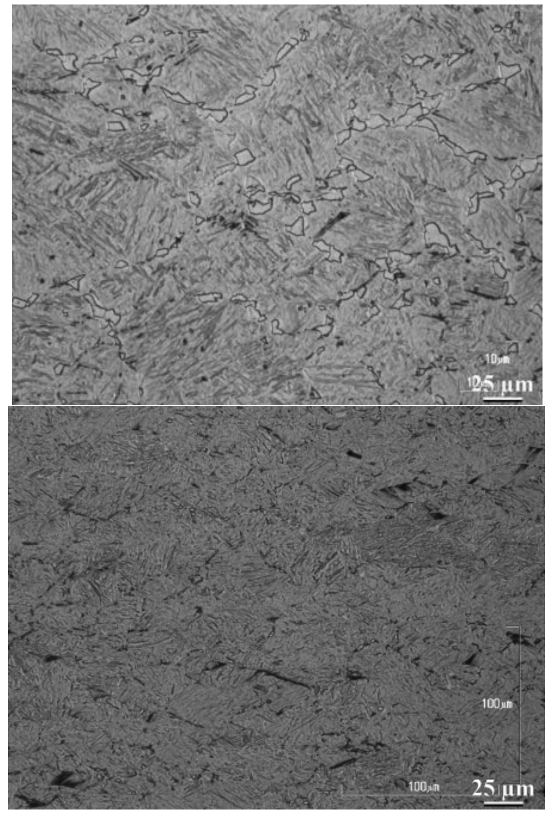

Figure 8. 0.5C-0.77Mn-1.90Al-V-Cr; a) FRT 850°C, b) FRT 900°C

3.3. Wear Properties

The results of the crushing pin-on-disk tests done for the rolled and quenched test materials and two commercial reference materials are presented in Figure 9. It is clear that higher hardness resulted in lower average mass loss. When comparing the results of the reference material Raex 450 and the laboratory steels with the smallest mass loss, it can be concluded that increasing the hardness from the 457 HV of Raex 450 to 570 HV or more can result in over 35% savings in mass loss.

It can also be seen that at a hardness of about 450 HV, the multiphase laboratory steel with 11% ferrite (0.4C-1.50Mn-2.10Al-V-850°C) suffered more mass loss than the fully martensitic reference material Raex 450 (457 HV). This suggests that for a given hardness, a ferritic-martensitic microstructure is not as abrasive resistant as a totally martensitic microstructure.

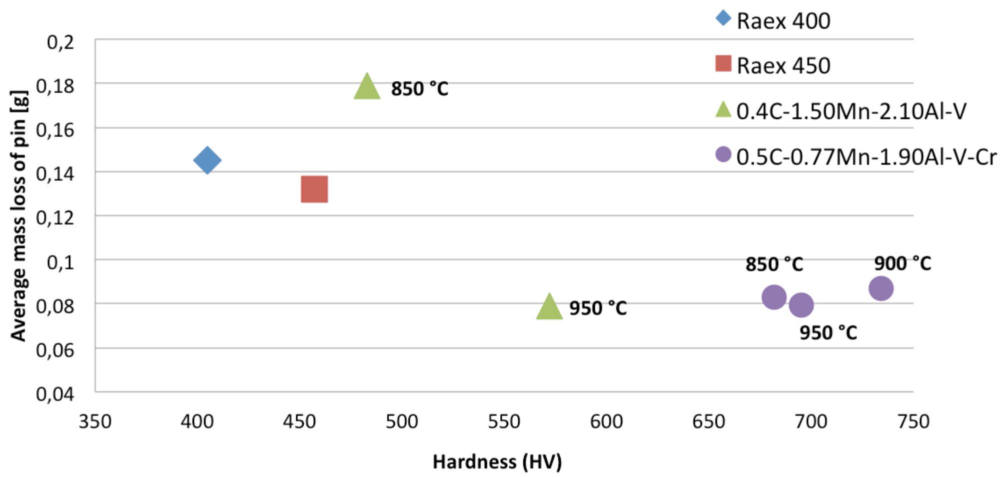

Figure 9. Results of the crushing pin-on-disk tests: mass loss vs. hardness

The huge difference in mass loss between the specimens with 11 and 1% ferrite made from 0.4C-1.50Mn-2.10Al-V also suggests that the dual-phase microstructure is not beneficial for abrasion resistance. This is contrary to the results reported by Jha et al.[3], who found that for a specific hardness level and composition, a ferritic-martensitic microstructure had better abrasion resistance than that of totally martensitic steel. They argue that dual-phase microstructure gives a good combination of hardness and toughness/ductility, which is more beneficial to abrasion resistance than merely a high hardness. Their studies concerned steels with lower hardness levels (252-451 HV), however. There were also major differences in testing methods between this work and the one used by Jha et al., which may explain the different observations. They used crushed silica sand (quartz) with particle size of 212-300 µm as abrasive. Quartz is much harder than granite and the particle size was much smaller than used in this work. They have also used the rubber wheel abrasion test described in ASTM G65-81. Hylkilä[14] has stated that when testing specimens hard enough using the rubber wheel abrasion test, the abrasive starts to penetrate into the rubber wheel, which causes the wear rate to increase with increasing hardness up to a certain level. This changes the wear from three-body abrasion to more like two-body abrasion, where the abrasive particles are rigidly attached to the counterbody[15]. Two-body abrasion tests are known to produce higher wear rates than three-body abrasion under comparable loading[15,16]. This could explain the lower wear rate obtained at a lower hardness reported by Jha et al.[3].

Based on the results presented in Figure 9, the abrasion resistance increases with the hardness as expected. However, further increase in hardness beyond about 600 HV does not show further improvement in abrasion resistance for the conditions studied here.

4. Summary and Conclusions

Physical simulation in form of Gleeble CCT diagrams was used for material selection and designing rolling experiments for obtaining dual-phase steels with hardness over 500 HV. Rolling experiments with direct quenching from varying final rolling temperatures were carried out. Abrasive wear resistance was determined using crushing pin-on-disc tests with granite and the results were compared to those from two commercially available reference materials. Martensitic and dual-phase steels with varying proportions of ferrite and martensite having hardness from 480 to 734 HV10 were obtained. The results indicated that increasing hardness from 450 HV of a fully martensitic steel to 570 HV or more can result in over 35% reduction in mass loss. However, under the present test conditions, the dual-phase steel with ferrite and martensite had poorer abrasion resistance than that of the totally martensitic reference material with the same hardness.

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