Evaluation of the potential of high Si high C steels nanostructured bainite for wear and fatigue applications.

T. Sourmail[1,a], F. G. Caballero[2], C. Garcia-Mateo[2], V. Smanio[1], C. Ziegler[3], M. Kuntz[4], R. Elvira[5], A. Leiro[6], E. Vuorinen[6], T. Teeri[7]

1) Ascometal-CREAS (research centre) Metallurgy, BP 70045 57301 Hagondange Cedex, France
2) Dept of Physical Metallurgy, National Centre for Metallurgical Research (CENIM-CSIC), Madrid, Spain
3) ALD Vacuum Technology GmbH, Wilhelm Rohn Str. 35, Hanau, Germany
4) Robert Bosch GmbH, Materials and Process Engineering, P.O. Box 300240, Stuttgart, Germany
5) Gerdau I+D, S. A. Barrio Ugarte, 48971 Basauri, Spain
6) Dept of Applied Physics and Mechanical Engineering, Lulea University of Technology, Sweden
7) Metso Minerals Oy, PO Box 237, 33101 Tampere – Finland

a) thomas.sourmail@ascometal.com

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Abstract

The present study is concerned with the potential of high carbon, high silicon steel grades isothermally transformed to bainite at low temperature (< 300 °C). A first part gives an overview of design principles allowing very high strength and ductility to be achieved, while minimising transformation duration. Wear and fatigue properties are then investigated for over ten variants of such material, manufactured in the laboratory or industrially. The results are discussed against published data. Tensile strength above 2 GPa are routinely achieved, with, in one case, an exceptional and unprecedented total elongation of over 20%. Bainite plate thickness and retained austenite content are shown to be important factors in controlling the yield strength, though additional, non negligible parameters remain to be quantified. Rolling-sliding wear performances are found to be exceptional, with as little as 1% of the specific wear rate of conventional bainitised 100Cr6. It is suggested that this results from the decomposition of retained austenite in the worn layer, which considerably increases hardness and presumaby introduces compressive residual stresses. Fatigue performance were slightly improved over 100Cr6 for one of the two industrially produced material, but significantly lower otherwise. Factors controlling fatigue resistance require further investigations.

Introduction

The past decade has seen increased research activity in the area of the so-called superbainite or nanostructured bainite [Caballero, 2002a; Garcia-Mateo, 2003a; Bhadeshia, 2004a; Soliman, 2007; Hodgson, 2011; Timokhina, 2011; Cruz Jr, 2012]. These microstructures are achieved through bainitising of high silicon steel grades with low Ms temperature, a process similar to that already applied to grades such as 100Cr6 for particular applications such as diesel injectors or large bearings. Whilst they were initially referred to as carbide free, there is today increasing evidence that cementite is present in the bainitic ferrite (as in conventional lower bainite). Nevertheless, in contrast with 100Cr6 or equivalent, high Si high C grades bainitised at low temperatures exhibit a significant amount of retained austenite.

Most of the published work to date has been dedicated to improving the kinetics of transformation and evaluating basic mechanical properties such as tensile strength and toughness [Garcia-Mateo, 2003a; Garcia-Mateo, 2005b; Caballero, 2007]. With little data available on complex properties such as fatigue and wear resistance, a joint research project was initiated between different partners, and partly funded by the European Commission. The aim of this project was to investigate such properties on grades specifically designed to provide bainite transformation within industrially acceptable times.

Different high silicon (1.5 wt%) grades were designed based on 0.6C and 1C wt% contents. Exceptional results were achieved in terms of tensile and wear properties. Fatigue properties were on par with existing material though no significant improvement could be obtained. The present paper provides an overview of these results, most of which are being published separately in greater details.

Grades design and manufacture, heat-treatments

A total of 10 grades were investigated around requirements corresponding to two potential industrial applications. The first corresponded to relatively small diameters components (approx. 30 mm diameter), the manufacturing of which was to be carried out using dry bainitising, a proprietary process whereby a steel is gas quenched from austenite to the bainitising temperature then transferred to a holding furnace. The second corresponded to a much larger component (approx. 400x700x200 mm^3) which was to be bainitised using a salt-bath.

The carbon content having been selected to achieve UTS of 1600 and 2000 MPa (0.6 and 1C wt% respectively), grades were designed using two basic principles. First, to accelerate bainite formation kinetics, alloying additions have to be kept to a minimum. Second, alloying additions need to be sufficient to avoid formation of any other transformation products than that expected from the isothermal holding at the bainitising temperature (for example pearlite). The optimisation was thus carried out by first estimating quantitatively the cooling rate of each target component in relevant conditions (gas quenching or salt bath), then tailoring the hardenability to the calculated cooling conditions while ensuring that the martensite start temperature Ms was sufficiently low to allow bainitising in the temperature range 200-250 °C. In both cases, the constraint on the Ms temperature was less severe than that for sufficient hardenability so that this criterion had no practital influence on the composition choices.

For economical reasons, Ni (for hardenability) and Co additions (for acceleration of bainite formation, [Aaronson, 1966; Garcia-Mateo, 2003a]) were avoided. High Al contents, which have also been used to accelerate bainite formation [Garcia-Mateo, 2003a], were also avoided as they are detrimental to steel cleanliness (which is critical for fatigue of high strength material). Finally, Mn additions were limited to approximately 1.5 wt% to keep segregation at acceptable levels.

The target compositions thus designed are shown in Table 1. Further details concerning the steel grade design are provided in [Garcia-Mateo, 2012x].

Steel	C	Si	Mn	Cr	Mo	Others
100Cr6*	1,00	0,25	0,30	1,50	-	-
1C	0,99	1,58	0,76	0,46	0,02	---
1C-Nb	1,02	1,53	0,76	0,46	0,02	0,02Nb
1C-Mo	1,01	1,51	0,82	0,46	0,10	---
1C-2.5Si	0,98	2,9	0,77	0,45	0,01	---
0.8C	0,90	1,65	0,79	0,48	0,01	---
0.6C	0,68	1,60	1,25	1,50	---	---
0.6C-2.5Cr	0,61	1,45	0,76	2,42	---	---
0.6C-Nb	0,64	1,6	1,27	1.50	0,01	0,04Nb
0.6C-Mo	0,58	1,58	1,25	1,43	0,02	---
1CSi-indus	0,99	2,47	0,74	0,97	0,03	
06CV-indus	0,67	1,67	1,31	1,73	0,15	0.12V

*Table 1 : Chemical composition of the steel grades investigated, in wt%; *typical composition for 100Cr6.*

Except for two steel grades, referred to as 1CSi-indus and 06CV-indus, all grades were manufactured in the laboratory, using vacuum induction melting to produce ingots of 20 or 35 kg. Once solidified, the ingots were reheated to 1200 °C (1150 °C for the 2.5% Si grades) and forged to bars of approximate diameter 40 mm. The bars thus forged were slowly cooled in a furnace to avoid cracking. After cooling, bars were heat-treated 2 h at 690 or 700 °C to further soften the material and ease machining of mechanical testing specimens. Industrial heats were manufactured industrially using electric arc furnaces. They were then hot-rolled to 60, 120 and 300 mm diameter bars for 06CV-indus, and to 35 mm bars for 1CSi-indus.

Metallography and retained austenite measurements

For laboratory examinations, sections of typically 10 mm thickness were taken and heat-treated either in salt-bath or using dry bainitising. They were prepared following standard metallography practice and examined in light optical microscopy and scanning electron microscopy (SEM) after etching in 4% Nital. Details of the manufacturing of mechanical testing specimen are given in later sections. High magnifications SEM micrographs were used for quantifying the bainite plate thickness t_β from the mean linear intercept $L=\pi t_\beta/2$ [Chang, 1995; Garcia-Mateo, 2003a]. Hardness measurements were carried out using Vickers indentation under 30 kg load unless otherwise specified.

Retained austenite measurements were carried out at two different laboratories. In a first case, a Siemens D5000 diffractometer was used with an unfiltered Co Kα radiation, a scanning speed of less than 0.3 °/min and operating at 40 kV / 30 mA. Four austenite peaks (111), (200), (220) and (311), and three ferrite peaks (002), (112), (022) were used. In a second case, a Brucker D8 was used, scanning a total of five peaks : (111), (200) and (220) in austenite and (110) and (200) in ferrite, under 12 different incident beam conditions to minimise texture effects. At the onset of the project, a round robin was carried out to verify that both methods gave identical results to within ± 1%.

Hardness and tensile properties

Results

Tensile specimens were machined from softened material, to within +0.5 mm of the final dimensions. Specimens were then heat-treated using dry bainitising (for 0.6-0.8C wt% materials) or salt bath (for 1C wt% grades). Austenitising temperature was set to 890 °C for 0.6 and 0.8C wt% materials and to 950 °C for 1C wt% materials, thus ensuring that carbon was fully in solution in all cases. To set appropriate isothermal holding durations, bainite kinetics were followed in a Bähr 805 dilatometer for a heat-treatment schedule identical to that later used for heat-treating of specimens. It was therefore ensured that all tests were carried out for fully transformed microstructures. Details of the TTT diagrams thus obtained are given in [Garcia-Mateo, 2012x; Garcia-Mateo, 2012xb; Sourmail, 2012]. Following heat-treatment, specimens were hard machined to the final dimensions. Care was taken to ensure that all specimens would be hard machined in identical conditions to achieve similar surface roughness. Typically, 3 to 5 specimens were used for each tensile test. This is critical as tensile testing at such strength level is notoriously sensitive to experimental conditions, so that demonstrating reproducibility is essential before discussing results.

Results are shown in Table 2. Many of the 1C wt% steels investigated are relatively close to the standard 100Cr6 grade, from which they differ mainly through their Si content. It is thus worthy to bear in mind that the 100Cr6 grade, when used after bainitisation, exhibits yield and tensile strengths of typically 2150 and 2350 MPa (resp.), with negligible uniform elongation, about 1.5% total elongation, and with approximately 2-3% retained austenite.

Condition	Heat-treatment		tβ / nm	Ret. A. / %	YS / MPa	UTS / MPa	UE / %	TE / %	Hardness (HV30kg)
	Tβ / °C	duration / h							
1C-250C	250	16	38	20	1834	2205	3,9	11,2	652
1CMo-250C	250	16	32	22	1852	2164	2,9	8,3	659
1CMo-220C	220	22	32	25	1883	-	0,3	0,3	707
1CMo-200C	200	64	21	20	2019	-	0,4	0,4	751
1CNb-220C	220	22	30	23	1798	-	0,9	0,9	714
1CNb-240C	240	16	28	21	1866	2278	4,9	7,2	691
1CSi-250C	250	16	39	33	1698	2068	11,6	21,3	625
1CSi-220C	220	22	28	36	1704	2287	7,4	7,4	650
06C-250C	250	12	60	12	1484	2023	7,8	14,2	591
06CMo-250C	250	12	43	20	1443	2022	7,0	19,1	602
06CNb-220C	220	22	60	27	1358	2234	5,1	5,1	634
06CNb-250C	250	12	65	18	1480	2017	7,6	19,1	595
06CCr-250C	250	12	43	19	1582	2030	7,8	14,2	589
08C-220C	220	40	35	16	1931	2329	3,2	4,1	710
08C-250C	250	22	37	18	1910	2213	3,0	11,9	659
08C-270C	270	7	36	24	1701	2036	4,4	12,6	615

Table 2 : Microstructural parameters for conditions whose tensile properties were investigated, together with results from tensile tests and hardness measurements. T_β is the bainitising temperature and t_β the plate thickness. Specimens for 1CMo-220C, 1CMo-200C and 1CNb-220C all ruptured during loading so that UTS cannot be measured.

Comparison with previous data

As reported in Table 2, UTS values were always in excess of 2 GPa. It is interesting to compare the results obtained above with previously reported data (Table 3).

Steel composition	Bainitising conditions	Plate thickness / nm	Hardness (HV)	YS and UTS if measured	Reference
0,98C-1,46Si-1,89Mn-0,29Mo-1,26Cr-0,09V	Austenitising at 1000 °C, bainitising 3-4 days at 250 °C	49	550	-	[Garcia-Mateo, 2003b]
0,80C-1,59Si-2,01Mn-0,24Mo-1,0Cr-1,51Co	Austenitising at 900 °C, bainitising at 250 °C for 14h	49	589	1,50 / 2,05 GPa	[Garcia-Mateo, 2005a]
0,80C-1,59Si-1,94Mn-1,33Cr-0,30Mo-0,11V	Austenitising at 1000 °C, bainitising 2 days at 250°C	50	580	2,5 GPa **compression**	[Caballero, 2002a]

Table 3 : Compilation of published data on similar low temperature bainitic microstructures.

A comparison of the typical transformation durations underlines the importance of tailoring the hardenability to the intended application and cooling method. Thus, while bainitising treatment of up to 4 days are required when heavy alloying additions are made (example 1 in Table 3), this time can be reduced, for the same bainitising temperature, to approximately 16 h.

Although strength in excess of 2.5 GPa has been reported, [Caballero, 2002a; Bhadeshia, 2005a; Bhadeshia, 2005b; Bhadeshia, 2005c; Bhadeshia, 2007a], this is only for compression tests, which cannot be compared to tensile properties. The present study has only been concerned with tensile strength, and led to similar strength levels as previously published (Table 3). Indeed, even 100Cr6 did not exhibit strength in excess of 2.4 Gpa.

Questioning the need for lower temperatures

Recent publications have been concerned with ever lower transformation temperatures (190 °C [Caballero, 2002a]; 200 °C [Bhadeshia, 2005b]; 125 °C, [Bhadeshia, 2005a]). However, the present results question the need for bainitising at such low temperatures. Indeed, neither the bainite plate thickness nor the UTS appear to be clearly increasing with decreasing temperatures below approximately 240 °C. Furthermore, as shown by a number of examples, there is a drastic reduction in ductility below a temperature of approximately 240 °C.

Figure 1 : Total tensile elongation for a variety of experimental grades bainitised at 220 or 250 °C. While UTS and YS were similar for both bainitising temperatures (Table 2), a systematic drop in ductility was identified when using lower temperatures.

This is further confirmed when considering Charpy impact test results, which were systematically in the range 10-20 J for bainitising temperatures above 240 °C, but around 5 J for 220 C [Sourmail, 2012].

Factors controlling the strength

Tensile strength in low temperature bainite has been related to retained austenite content and bainite plate thickness [Garcia-Mateo, 2005a]. A useful parameter is thus Ret. A./t_β, as the strength should approximately vary linearly with the retained austenite content and the inverse of the plate thickness.

Figure 2 : Yield strength of experimental alloys as a function of the retained austenite content and inverse of the bainite plate thickness.

As shown in Figure 2, there is a clear correlation between the selected parameter and the yield strength. Differences of up to 200 MPa between the mean predicted value and actual result are however beyond measurement error and represent 1/3 of the maximum YS increase from the "softest" to the hardest material. Isolating grades of various carbon content does not improve the correlation. Therefore, while retained austenite and bainite plate thickness do constitute the first order parameter in controlling the strength, additional, non negligible mechanisms are at play, which remain to be clarified. These may include carbon segregation along dislocation, clustering, *etc.* [Caballero, 2007].

Exceptional combination of strength and ductility

As shown in Table 2, the 1CSi grade (~3 wt% Si) bainitised at 250 °C achieved an unprecedented uniform elongation of nearly 12%, and total elongation of nearly 22%, for a UTS of nearly 2.1 GPa. It is tempting to attribute this result to the amount of retained austenite (33%), which is significantly above that found in other 1C wt% materials. However, as discussed in the previous section, bainitising at

220 instead of 250 °C led to comparable amounts of retained austenite, though considerably lower ductility. It is thus likely that retained austenite stability and not simply its amount, is key to control ductility. This is further discussed in [Garcia-Mateo, 2012a]. The question of the influence of Si on the amount of retained austenite is discussed elsewhere [Sourmail, 2012b]. Indeed, the commonly stated effect of Si is that of suppressing cementite from contents around1.5-2 wt%, so that no further evolution is expected beyond these additions.

Wear resistance

Introduction

Wear properties of bainitic structures have essentially been studied in the context of rail steels. Rail steels have long been manufactured using pearlitic steels, the wear resistance of which increases with their hardness [Clayton, 1980]. As bainitic microstructures offer much higher hardnesses for lower carbon contents, much research has been devoted to replacing pearlitic steels with bainitic ones. This is because the latter were expected to offer improved wear resistance with better weldability.

The early work on wear resistance of bainitic steels for rails produced disappointing results both in the laboratory and on the tracks [Clayton, 1992; Clayton, 1996], with reported rolling/sliding wear rates up to ten times greater than those of pearlitic steels. As explained below however, these pessimistic early conclusions came from an oversimplified understanding of what constitutes bainite. As pointed out by Clayton and Jin [Clayton,1996], four types of bainitic microstructures should be distinguished: upper and lower bainite, carbide-free bainite and globular (presumably granular) bainite. Not all bainites perform equally during wear tests.

The work of Clayton [Clayton, 1980] and Clayton et al. [Clayton, 1987] suggested that bainitic microstructures did not perform significantly better than pearlitic ones in dry rolling/sliding tests except at relatively low hardness. In further experiments on low carbon bainitic steels however, it was found that an 0.04 %C bainitic steel outperformed harder pearlitic steels in rolling-sliding wear tests [Devanathan, 1991]. Later, Clayton and Devanathan [Clayton, 1992] investigated the wear of 50 HRC bainite formed by isothermal treatment at 290 °C and reported a far greater wear resistance than for a pearlitic microstructure. Interestingly, the same material transformed to a bainitic microstructure of hardness equal to that of the pearlitic one did not perform as well as the latter.

Figure 3 : Wear rate of different bainitic steels as a function of hardness, after [Clayton, 1996].

These results, in contrast with previous conclusions by other authors [Kalousek, 1985], suggested that much was to be gained from an increase in bainite hardness (for lower bainite at least). Additional results confirmed the excellent potential of bainitic microstructures, as Clayton and Jin [Clayton, 1987; Clayton, 1996] reported exceptional wear resistance in high silicon cast bainitic steels (Figure 3) and confirmed that carbide free structures gave the best wear resistance (rolling/sliding).

From their results, the authors clearly anticipated that increasing the hardness while retaining the carbide free microstructure should enable outstanding wear resistance.

Since then, there are numerous additional reports of the good sliding wear performance of bainitic steels containing retained austenite, for example, Shipway et al [Shipway,1997], or Vuorinen et al [Vuorinen,2006], who reported excellent results on spring steels (0.6C, 1.9Si) isothermally transformed to bainite, both in dry sliding wear tests and in service (knives in a hammer mill used for cutting saw dust).

The excellent results obtained with high silicon bainitic structures are generally attributed to the combination of high strength and toughness resulting from the ultrafine microstructure and absence of coarse carbides.

Testing procedure

In order to replicate rolling/sliding wear conditions, two specimens (discs of 45 mm diameter, 10 mm thickness) of the same composition and similar surface roughness were paired and mounted on a UTM 2000 twin disc machine for each test. The wear tests were run in dry conditions at 100 rpm and 95 rpm respectively, creating a 5 % slip between the two discs. Using such a high slip percentage is needed in order to maintain reasonable test durations, while still obtaining measurable weight losses. The applied load was set to 300 N for all tests and running time was 5 hours, all tests were carried out at room temperature and with approximately 25% humidity. The UTM 2000 can be used for solid state friction without lubrication, or boundary and mixed conditions with liquid lubricants, which allows for tests of both lubricants and materials. In dry conditions, the linear wear of both specimens as well as the friction coefficient is measured continuously. The applied load is measured with a force sensor on the load lever and the friction coefficient is calculated by measuring the traction force between the discs, using a torque sensor. The surface roughness of the samples was analysed by vertical optical interferometry before testing to ensure absence of influence from the initial surface condition.

The quantitative analysis of the wear performance was then determined by measuring mass loss, wear rate and determining the specific wear rate (K), which is calculated as the ratio of the mass loss for both discs divided by (sliding distance x load). Further details can be found in [Leiro, 2012] and [Sourmail, 2012].

Grade	Microstructure	Elements / wt%					
		C	Si	Mn	Cr	Ni	Mo
40CrMo8	Bainite	0,39	0,50	0,90	2,02		0,43
60CrSiMo9	Pearlite	0,61	0,93	0,62	2,19		0,45
60SiCr7	Varied	0,61	1,72	0,75	0,35	0,12	0,04
100Cr6	Varied	1,00	0,30	0,30	1,50		
42CrMo4	Varied	0,42	0,25	0,75	1,05		0,20

Table 4 : Composition of additional references for wear testing.

Most materials presented in Table 1 were tested. Specimen references include the bainitising temperature (for example 1CSi-250C refers to grade 1CSi bainitised at 250 °C). In order to provide comparison to additional standard material, results are included for tests carried out in identical conditions, for the materials detailed in Table 4 and Table 5.

Grade	Reference	Heat-treatment	Microstructure
40CrSi8	**40CrSi8**	undisclosed, hardness about 400 HV	"Conventional" bainite
60CrSi9	**60CrSI9**	undisclosed, hardness about 300 HV	Pearlite
60SiCr7	**60SiCr7-250**	Austenitising 860°C, bainitising at 250 °C	Carbide free bainite + retained austenite + tempered martensite
	60SiCr7-300	Austenitising 860°C, bainitising at 300 °C	Carbide free bainite + retained austenite
	60SiCr7-350	Austenitising 860°C, bainitising at 350 °C	Carbide free bainite + retained austenite
	60SiCr7-N	Austenitising 860 °C, natural cooling (normalising)	Pearlite
100Cr6	**100Cr6-std**	Standard (austenitising ~860°C, bainitising 250 °C)	Conventional lower bainite + coarse globular carbides
	100Cr6-950C-250C	Austenitising 950 °C, bainitising 250 °C	Conventional lower bainite
	100Cr6-950C-300C	Austenitising 950 °C, bainitising 300 °C	Conventional lower bainite
42CrMo4	**42CrMo4-B**	Austenitising 860°C, bainitising at 350 °C	Bainite
	42CrMo4-B	Austenitising 860°C, bainitising at 400 °C	Bainite
	42CrMo4-QT	Quenched and tempered	Tempered martensite

Table 5 : Heat-treatments for the additional wear test references.

Results and discussion

Results are summarised in Figure 4. As can be seen, with only the exception of the 06CV grade bainitised at higher temperature (280 °C), all measured specific wear rates were below 10^{-4} $mm^3.N^{-1}.m^{-1}$. Specific wear rates measured on industrially manufactured material (1CSi-ind-220, 06CV-220C) were as little as 1% of that measured on conventional 100Cr6 bainitised in similar conditions.

Figure 4 : Compilation of wear test results for designed low temperature bainitic steels and for various references.

Results thus confirmed the exceptional potential of low temperature high Si bainitic steels for wear applications. In order to clarify the origin of the differences in behaviour, retained austenite measurements were carried out underneath the worn surface, at 25 and 200 µm. For this, the surface of the specimens was electrolytically polished until reaching the desired depth. In addition, microhardness measurements were carried out underneath the worn surface to quantify the hardness increase and its extension in depth. As shown in Table 6, an evolution can be identified from 100Cr6, which exhibits virtually no retained austenite and limited surface hardening, to 1CSi-220C which exhibits a significant amount of retained austenite (30%) and considerable but very localised surface hardening over depth corresponding to those for which destabilisation of retained austenite is observed.

Reference	$K / m^3/(N.m)$	R. A. / %			Surface hardness after testing
		surface	@ 25 µm	@ 200µm	
1CSi-indus-220C	8,6E-06	14	29	30	1150
06CV-indus-220C	1,2E-05	-	16	32	1247
06C-250C	9,9E-05	-	7	17	954
100Cr6-300C	2,3E-04	-	-	3,7	935

Table 6 : Evolution of retained austenite and microhardness underneath the worn surface for 3 of the designed alloys and for standard 100Cr6 bainitised at 300 °C.

Thus, the exceptional wear resistance of the high Si, low temperature bainitic steels investigated can presumably be attributed to the transformation of retained austenite into martensite of very high hardness. This martensite would be expected to be significantly harder in 1CSi-indus than 06CV-indus, though the very limited spatial extent of the destabilisation certainly led to an underestimation of the local hardness in the case of 1CSi-indus.

Fatigue properties

Introduction

Fatigue is perhaps the least investigated property of low temperature bainites. Peet et al [Peet, 2012]. have discussed early results of fatigue on a Fe-0.80C-1.59Si-2Mn-1Cr-0.24Mo-1.5Co (wt%) steel bainitised for 10 days at 200 °C to achieve a hardness of 680 HV. A total of 10 fatigue tests were carried out by the authors with a maximum number of cycles of 10^5 and a load ratio R=0.1.From the results, the authors attempted to estimate the fatigue strength at 10^7 cycles by drawing a straight line from the UTS at 0 cycles, through data points at 10^4-10^5 cycles, up to 10^7 cycles where they estimated the corresponding stress. Not only is the method not valid because S-N curves exhibit slope changes around 10^2-10^3 and 10^5-10^6 cycles [Lieurade, 2000], but also, there is no recognised method to extrapolate data from 10^4-10^5 to 10^7 cycles. It is thus impossible, in this case, to draw any conclusions as to the performance of the material investigated.

Additional results have recently been provided for a similar material 0.83C-0.81Cr-1.56Si-1.37Mn-0.87W-1.44Al (wt%) [Yang, 2012]. Fatigue strength was estimated to 1147 MPa for tests carried out on smooth specimens at R=0.1. The authors discuss their results against the well known threshold of about 400HV below which sensitivity to cleanliness is negligible and fatigue strength is well approximated by UTS/2, and beyond which cleanliness dominates so that fatigue strength of UTS/2 are seldom if ever achieved. However, this is valid for a load ratio of R=-1, and the authors fail to correct their estimated fatigue strength to account for their different R ratio. Using the experimental results published by Yang *et al*. [Yang, 2012], the present authors have estimated a fatigue strength of approximately 790 MPa at 10^7 cycles for R=-1. This was done using a simple Goodman-Smith diagram [Bathias, 1997]. As shown in Figure 5, the results thus corrected lie well within the dispersion band for the zero mean stress fatigue resistance of quenched and tempered steels.

Figure 5 : Dispersion band for the 10^7 cycles fatigue resistance (R=-1) of various quenched and tempered materials, after [ASM, 1961]. Estimation for R=-1 based on results [Yang, 2012] have been added.

The above discussion clearly indicates that there is little reliable data available on the fatigue properties of nanostructured bainite, and so far no evidence of behaviour significantly different from other microstructures, as the estimations of fatigue limit for low temperature lies well within the typical dispersion band for quenched and tempered microstructures.

Testing procedure

In the present evaluation of fatigue performance of low temperature high carbon bainitic steels, notched specimens were used instead of smooth ones. Indeed, most of the applications for which such materials could be utilised exhibit stress concentrators, so that the fatigue performance in notched conditions is of higher relevance. Whilst results are more difficult to interpret or compare to the vast amount of published data for smooth specimens, comparisons remain possible between materials. Also, notched fatigue may be more sensitive to the presence of retained austenite as the stress concentration under the notch may enhance its decomposition.

In a first series of tests, notched rotation bending specimens (Kt=1.6) were manufactured from a number of experimental materials and from both industrial grades (Table 1). Rotation bending was selected because in such a procedure, stresses are mostly located near the surface, thus reducing the volume sampled and the risk of initiation on non-metallic inclusions (cleanliness of laboratory casts cannot be controlled during elaboration). With the addition of a notch, stresses are also localised along the specimen length thus further reducing the sampled volume. In a second series, notched tension-tension specimens (Kt=2) were manufactured from industrial casts 1CSi-indus and 06CV-indus and tested with a load ratio R=0.1.

All specimens were pre-machined after annealing the materials (typically 2 h at 690 °C), then heat-treated in conditions identical to those discussed earlier (Table 2). Final hard-machining was carried out on heat-treated specimens to achieve the required dimensions. Further details are available in [Sourmail, 2012] and will be the subject of a coming publication.

Results and discussion

Results of rotation bending tests are shown in Table 7 and those obtained in tension-tension conditions, in Table 8.

Material and bainitising temperature / °C	UTS / MPa	Sa,50% at 10^7 cycles / MPa for R=-1, Kt=1,6 by rotation bending
06C-250	2023	665
06CV-indus-250	2003	690
06CV-indus-270	1822	675
08C-270	2036	440
1CNb-220	2073	430
1CSi-indus-220	2224	550
1CSi-indus-250	2072	535

Table 7 : Notched (Kt=1.6) rotation bending fatigue strength at 10^7 cycles for different materials manufactured in the laboratory or industrially.

	06CV-250	06CV-270	1CSi-220	1CSi-250	100Cr6 [Linkewitz, 1998]	50CrMo4 [Schmidt, 2012]
UTS / MPa	2003	1822	2224	2072	2350	1180
$S_{a,50\%}$ at N = 10^7 cycles / MPa	350	330	205	245	355	305

Table 8 : Notched (Kt=2) tension-tension (R=0.1) fatigue strength at 10^7 cycles for industrially produced nanostructured bainite, and for two references tested in identical conditions.

Because references were not available in the exact testing conditions used for the rotation bending fatigue, results can only be compared with one another. A comparison of the fatigue strength achieved on 0.6C wt% and higher carbon material suggests a significant difference between these two types of grades. This is further supported by the fact that both the laboratory and industrially produced 0.6C wt% exhibited similar strength, and will be further discussed below.

Notched tension-tension performances were compared to those obtained on two published references: bainitised 100Cr6 with a UTS of 2350 MPa [Linkewitz, 1998], and quenched and tempered 50CrMo4 with a UTS of 1180 Mpa [Schmidt, 2012]. The 100Cr6 used for comparison was bearing quality material, that is, with the highest cleanliness achievable with present standard production techniques. As shown in Table 8, results achieved on the 1CSi material once again appear to be well below those obtained on lower carbon content material and on references. The reason for this may lie in lower cleanliness of the experimental material but also in the use of a high austenitising temperature. Indeed, as noted earlier, all the high carbon material was austenitised at 950 °C instead of 890 °C for lower carbon content grades. The reference 100Cr6 was also bainitised in standard industrial conditions, which implies austenitising temperatures around 860 °C as full carbon dissolution is not sought. The higher temperature may have led to larger prior austenite grain size which is known to be detrimental to fatigue strength. This is the object of further ongoing investigations. On the contrary, the 06CV-indus material was shown to exhibit similar fatigue strength as that of 100Cr6, in spite of a somewhat lower UTS. In addition, examination of the failed specimens revealed systematic initiations on large titanium carbonitrides. The latter are present because of a titanium content of 0.003 wt%, which is sufficient to cause formation of Ti(C,N) in the liquid when casting large ingots. Ti content down to less than 0.001 wt% can routinely be achieved in high cleanliness bearing steels and there is thus potential to exceed the performances of 100Cr6.

Conclusions

The application of low temperature bainitising heat-treatment to high carbon, high silicon steels has led to a new class of materials, which exhibit nanostructured ferrite plates together with retained austenite in significant quantities.

A first point addressed in the present work was that of transformation kinetics. It was shown that adaptation of the hardenability to the minimum requirement of the manufacturing process allowed to achieve transformation over acceptable durations, without the need for expensive elements such as Co or elements detrimental to cleanliness such as Al.

When transformed at temperatures below approximately 250 °C, such materials can lead to tensile strength well in excess of 2GPa, while retaining ductility. Similar strength levels have routinely been achieved on the standard 100Cr6 grade, either by low temperature bainitising or quench and tempering, though neither heat-treatment lead to appreciable ductility. It is clear that the presence of retained austenite in the presently investigated microstructure is the explanation for improvement in ductility. Indeed, the present work has led to the discovery of a unique combination of strength and elongation in very high Si materials for which a UTS above 2GPa was achieved in combination with over 30 % retained austenite [Garcia-Mateo, 2012a]. It was also shown that, as far as tensile properties and ductility are concerned, seeking the lowest possible temperature was possibly counterproductive as strength only moderately improved with decreasing temperature, whilst ductility decreased significantly.

As expected from extrapolation of earlier discussion of wear resistance, rolling/sliding wear of bainitised high carbon, high silicon steels was found to be considerably improved over standard lower bainite or quenched and tempered microstructures. In the context of wear, decreasing the bainitising temperature appeared to be beneficial, most likely as a result of increased hardness and because ductility has no influence. Specific wear rate of the hardest material tested in the course of the present project was as little as 1% of that of 100Cr6 bainitised to a UTS of 2350 MPa. Examinations of the worn surfaces suggest that this is related to destabilization of retained austenite underneath the worn surface, which results in a dramatic increase in hardness, compared to material with no or little retained austenite where this increase is less pronounced. This process will also most likely introduce compressive stresses which will inhibit propagation of the damage.

Fatigue performance was investigated in notched conditions. Published data, as with most of the results achieved in the present study, do not indicate particular advantages of this microstructure in comparison to conventional lower bainite (bainitised 100Cr6) or quenched and tempered materials. In one case however, results suggested that performances may exceed those of bearing quality bainitised 100Cr6. However, this potential increase remains to be quantified. Further work is thus necessary in order to determine whether nanostructured high Si bainitic steels may offer benefits in terms of fatigue performance.

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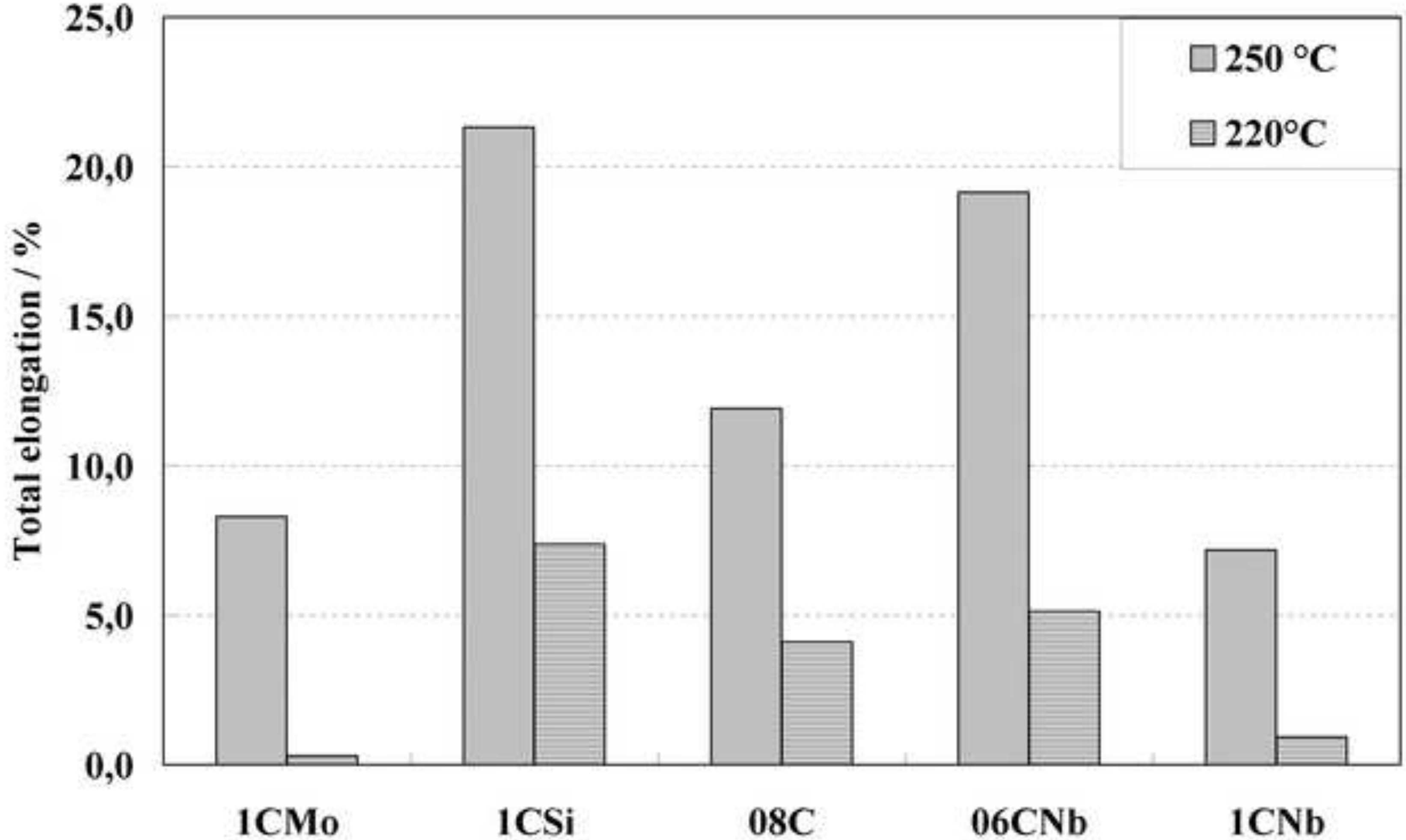
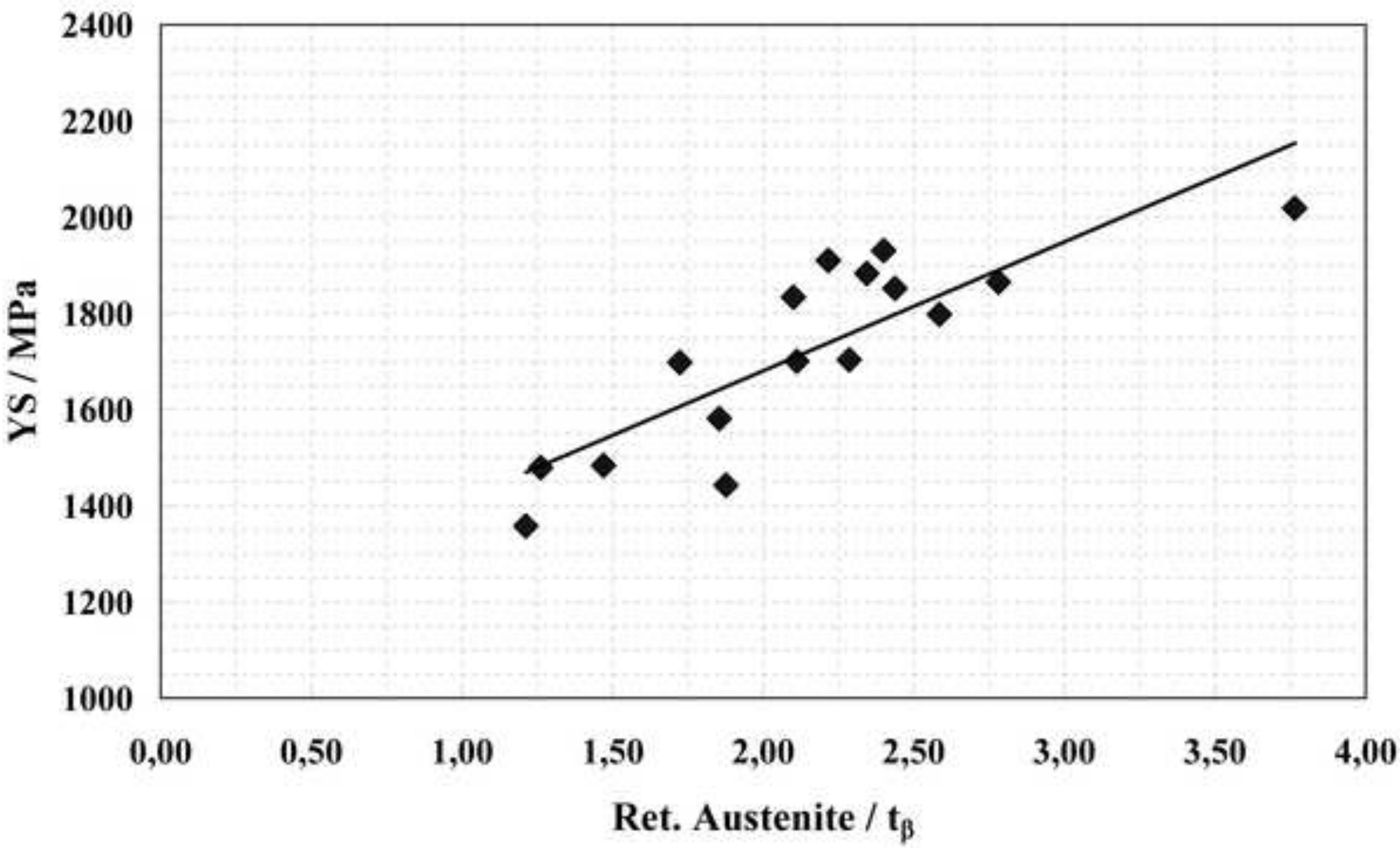
Figure 1



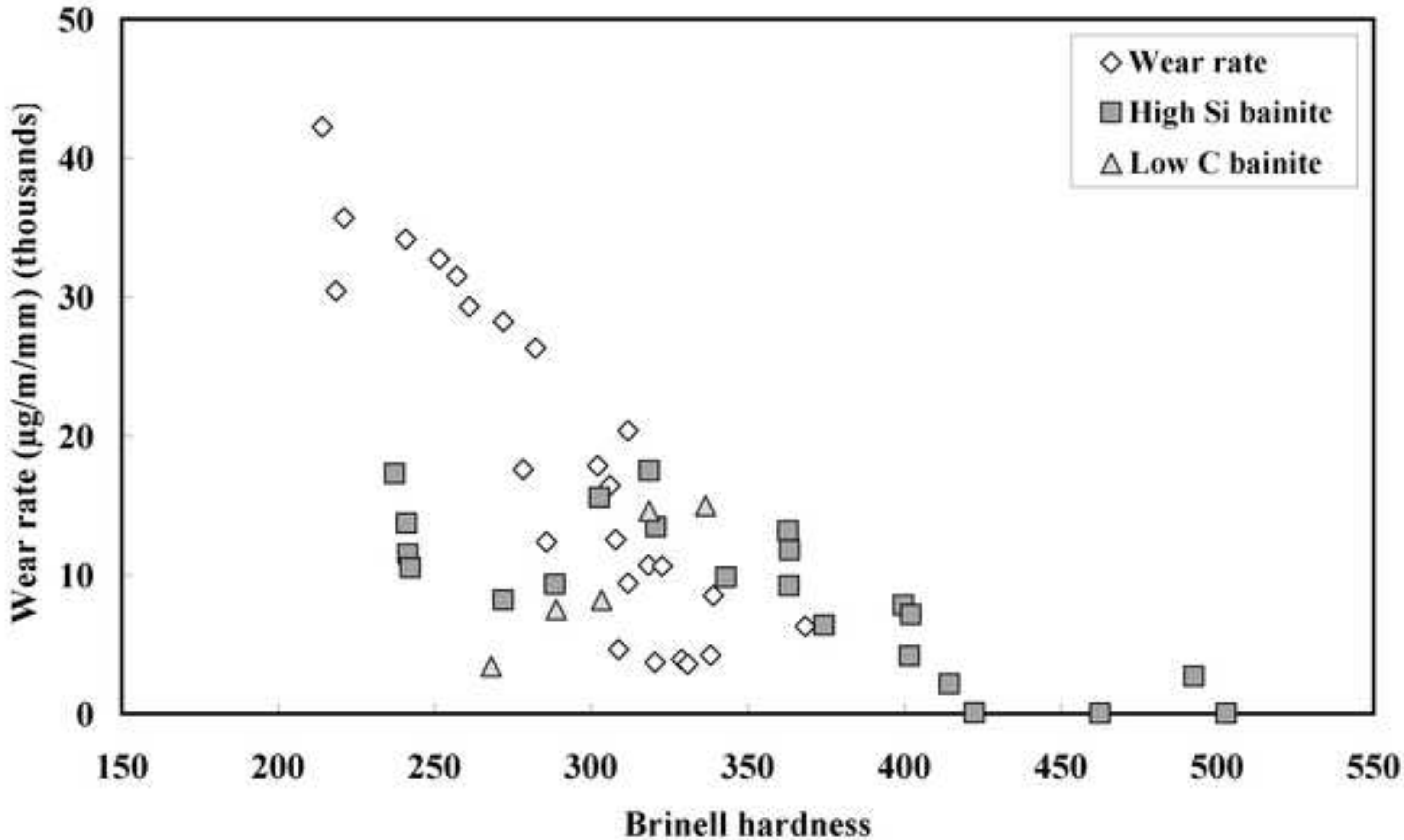
250 °C
220°C
25,0
20,0
15,0
10,0
5,0
0,0
Total elongation / %
1CMo
1CSi
08C
06CNb
1CNb

Figure 2



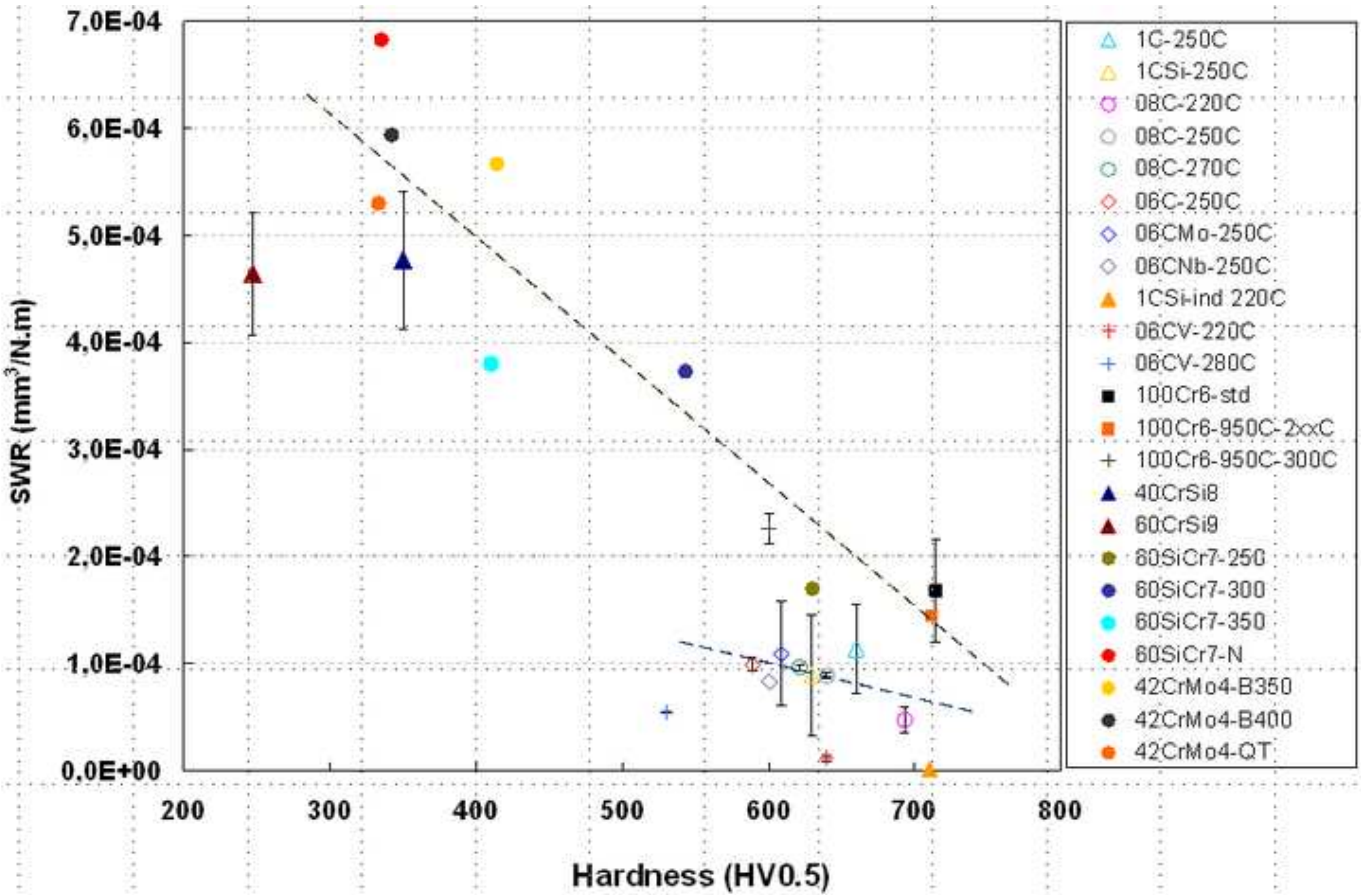
YS / MPa
2400
2200
2000
1800
1600
1400
1200
1000
0,00
0,50
1,00
1,50
2,00
2,50
3,00
3,50
4,00
Ret. Austenite / t_β

Figure 3



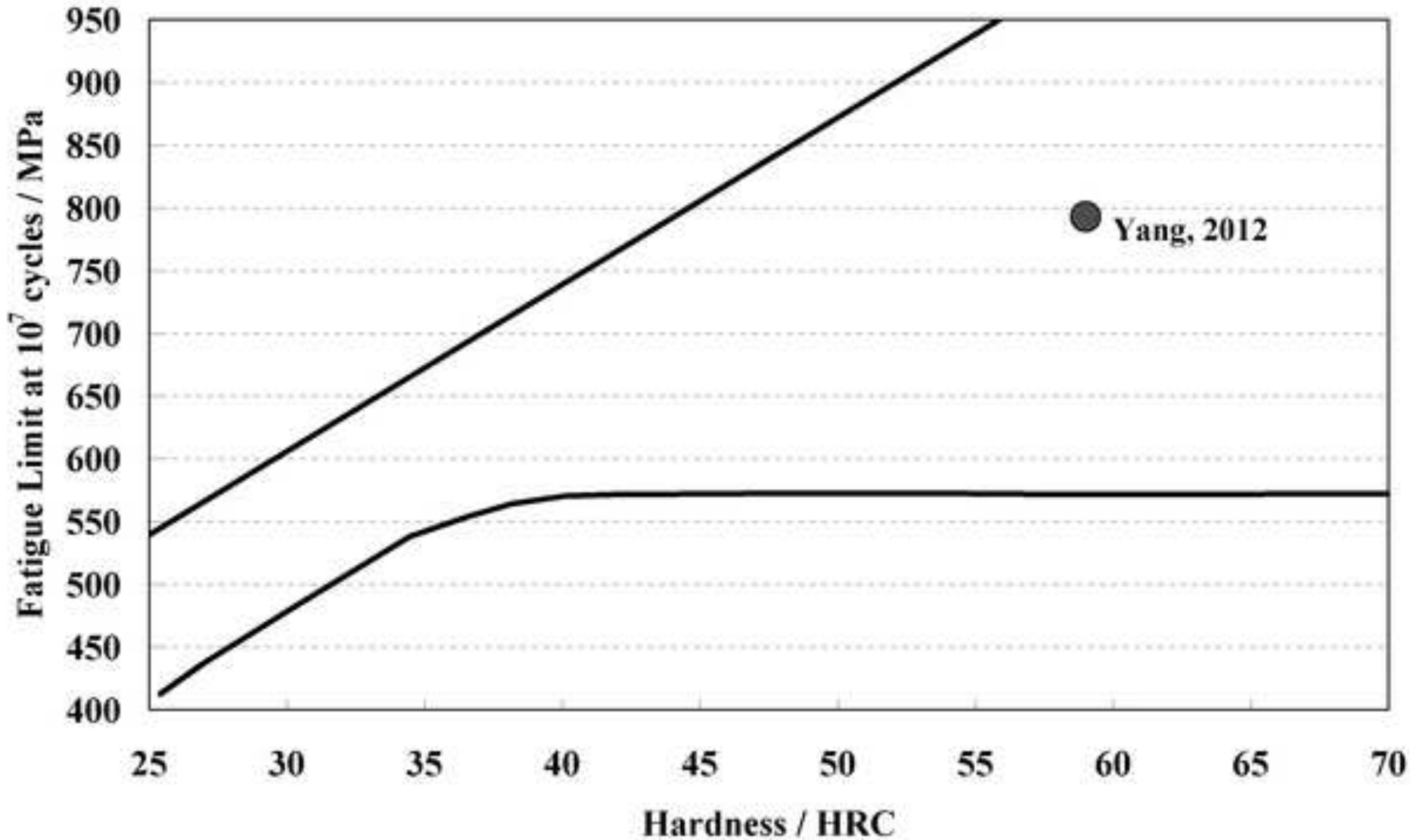
Wear rate
High Si bainite
Low C bainite
Wear rate (µg/m/mm) (thousands)
Brinell hardness
0
10
20
30
40
50
150
200
250
300
350
400
450
500
550

Figure 4



7,0E-04
6,0E-04
5,0E-04
4,0E-04
3,0E-04
2,0E-04
1,0E-04
0,0E+00
SWR (mm^3/N.m)
200
300
400
500
600
700
800
Hardness (HV0.5)
1C-250C
1CSi-250C
08C-220C
08C-250C
08C-270C
06C-250C
06CMo-250C
06CNb-250C
1CSi-ind 220C
06CV-220C
06CV-280C
100Cr6-std
100Cr6-950C-2xxC
100Cr6-950C-300C
40CrSi8
60CrSi9
60SiCr7-250
60SiCr7-300
60SiCr7-350
60SiCr7-N
42CrMo4-B350
42CrMo4-B400
42CrMo4-QT

Figure 5



950
900
850
800
750
700
650
600
550
500
450
400
Fatigue Limit at 10^7 cycles / MPa
Yang, 2012
25
30
35
40
45
50
55
60
65
70
Hardness / HRC