Journal of Achievements in Materials and Manufacturing Engineering
VOLUME 15
ISSUE 1-2
March-April
2006

Engineering of forged products of microalloyed constructional steels

J. Adamczyk, M. Opiela*
Division of Construction and Special Materials Engineering,
Institute of Engineering Materials and Biomaterials, Silesian University of Technology,
ul. Konarskiego 18a, 44-100 Gliwice, Poland
* Corresponding author: E-mail address: marek.opiela@polsl..pl

Received 15.11.2005; accepted in revised form 15.02.2006

Manufacturing and processing

ABSTRACT

Purpose: Effect of the thermo-mechanical treatment conditions on the structure and mechanical properties of the forged elements of constructional C-Mn steels with Ti, V, B and N microadditions.

Design/methodology/approach: Metallography, electron microscopy, tensile test, hardness measurements, hardenability calculations, Charpy-V tests have been used.

Findings: The thermo-mechanical treatment allows to obtain the fine-grained austenite structure during hot plastic deformation, and gives forged elements obtaining: yield point $R_{p0,2}$ over 690 MPa, UTS over 770 MPa, hardness 220 up to 250 HB and breaking energy KV over 180J after high tempering.

Research limitations/implications: It is predicted TEM investigations on structure of the forged elements after thermo-mechanical treatment.

Practical implications: Investigations carried out showed full usability of micro-alloyed steels for producing forged machine parts with the high strength and cracking resistance, using the energy-saving thermo-mechanical treatment method.

Originality/value: Production conditions of energy-saving thermo-mechanical treatment of forged elements of HSLA constructional steels – with the diversified hardenability, were presented.

Keywords: Thermo-mechanical treatment; Microalloyed steels; Forged elements; Mechanical properties; Cracking resistance

1. Introduction

Die forging is generally used as a manufacturing method of machine parts from non-alloy and alloy steels. Forged elements from non-alloy steels are usually normalized to improve their properties thanks to grain refinement. Forged elements of alloy steels have to be soft annealed or tempered in high temperatures, before their mechanical treatment and in ready state quenched and tempered in high temperatures with dimension correction by machining - especially grinding and finishing.

Possibility of cost reduction during producing forged elements because of elimination or reduction heat treatment, to tempering or artificial ageing only have appeared with developing micro-alloyed steels HSLA type. This kind of steels consists up to 0,5 % of C and up to 2% of Mn and Nb, Ti and V in quantity up to 0,1% and sometimes also Zr and small content of N and up to 0,005% B which increase the hardenability of these steels. Hot working of these steels in well chosen conditions allows to create fine-grained structure with dispersive particles of MX interstitial phases (M – Nb, Ti and V; X – N and C), precipitated in plastically deformed austenite. Those particles are limiting a grain growth of recrystallized γ phase. Fine-grained structure (as a product of transformined austenite) which occures in adjustable cooling conditions and precipitation hardening (follows from presence of MX phase particles) are deciding about considerable increasing of strength properties with preserving a good cracking resistance of manufactured products. That problems, widely described in many publications and also in proceedings of many scientific conferences, are also shown in papers [1-4].

Metallic micro-alloys introduced to the steel and also B have a strong affinity to oxygen and nitrogen. Besides that Ti and Zr have also affinity to sulfur. This means that the liquid steel, just before introducing micro-alloys, should be well deoxidided and desulphurized. Because of that steel heated in oxygen-blown converter or in electric arc furnace have to be treated in the drawing of a ladle furnace in conditions which guarantee energetic course of chemical reaction with liquid base slag (which is on the top of liquid metal) and powder substance blown to the liquid metal (which are needed for deoxidizing, desulphurization and modification of non-metallic inclusions) and alloying elements and micro-additives (without moisture and crystalic water thanks to theirs roasting). Gases used to blow inside these additives (argon or nitrogen) making intensive mixing of bath and homogenizing of chemical composition of the steel. The final stage of ladle treatment is vacuum deoxidizing of the bath and continuous casting in argon atmosphere, protecting steel from secondary oxidizing and nitriding. Ingot intersection is chosen taking to account intersection of forge charge with required plastic processing of ready forging .

Processing of forge elements with high mechanical properties from micro-alloyed HSLA steels require well chosen plastic treatment conditions accommodated to kind and kinetics of dissolution (precipitation) MX interstitial phases (from introduced to the steel microalloys) in austenite described by equation:

$$\log[M][X] = B - A/T, \qquad (1)$$

where:

- [M] and [X] - weight concentration of metallic and metalloid micro-additives dissolved in steel austenite in T temperature,
- A and B - constants.

Value of constants A and B in equation (1) for chosen interstitial phases and BN phases are presented in table 1.

Table 1.
Values of the constants A and B in the equation (1) for selected carbides and nitrides [1, 2]

	MX							
cons.	AlN	VC	VN	TiC	TiN	NbC	NbN	BN
A	7184	7840	9500	10745	8000	7290	8500	13970
B	1,79	3,02	6,72	5,33	0,32	3,04	2,8	5,24

In case of using boron, it is needed to introduce Ti to the bath in amount necessary to bind a nitrogen from the steel to TiN. This operation prevents possibilities of fixing the nitrogen to stable nitride BN type. Plastic deformation of HSLA steels especially the final stage is provided in a range of temperatures related with precipitation processes of MX phases in austenite. The important think is that in the temperature of the end of plastic deformation all micro-additives introduced to the steel have to be chemically bounded to MX phases. In these conditions process of the plastic deformation is controlled by dynamic recovery.

Technical - economical aspects are responsible for this, that contemporary predominant part of die forged elements for needs of: motorization industry, maining and agriculture machines and others are done from micro-alloys machine-steels with ferritic-pearlitic structure. These steels contain up to 0,5 % of C, up to 0,8% of Si and micro-additives of Nb, Ti, V and N (table 2).

The main condition to obtain desirable mechanical properties of forged elements is good choise of processing parameters (charge heating temperature and plastic processing temperature – deformation distribution and the strain speed are very difficult to be controlled) [3]. Charge heating temperature should not totally dissolved MX phases in the solid solution because this could cause superfluous grow of grain size. High speed deformation and short brakes between consecutive deformation stages during forging, limiting the effective run of static recrystallization - responsible for refining austenite grain size. Although transformation $\gamma \rightarrow \alpha$ of plastically deformed austenite (both coarse-and fine-grained) is starting at the grain boundary and strip deformation, but in case of coarse - grained γ phase this not assure to obtain sufficient refine structure and expected mechanical properties of forged elements. Forging elements prepared in these conditions, cooled down from end of plastic processing temperature on a fresh air have high strength (thanks to strong precipitation processes, and to high volume of pearlite) but small cracking resistance [5].

From data putted in table 2 it is obvious that elements from micro-alloys steels forged in adjust conditions have considerably less cracking resistance in comparison with one made from steels with alloying elements for toughening. Because of this sometimes theirs application are limited due to standard requirement and allow regulations. However, they have higher fatigue resistance, mainly good machinability and they do not need a heat treatment.

Higher strength properties (especially cracking resistance) in comparison with forging elements with ferritic-pearlitic structure have elements, forged using thermo-mechanical treatment from low-alloy steels for toughening, with microadditions of Ti, Nb and V and also N and B. This method is based on steel plastic deformation in adjusted forging conditions with usual or isothermal quenching forged elements directly from end forging temperature advantageously after $t_{0,5}$ time pass, which is needed to obtain 50% volume of statically recrystallized austenite. Direct common quenching after passing $t_{0,5}$ time eliminate heat treatment of forged elements only to high tempering. However, isothermal quenching eliminates completely expensive toughening (fig. 1).

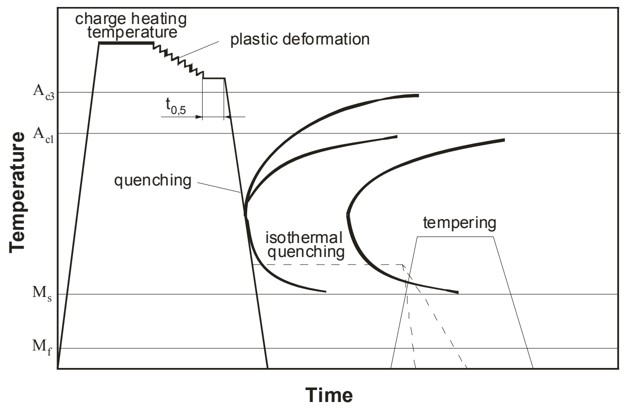

Fig. 1. The scheme of production of forge element from HSLA steels using thermo-mechanical treatment [3]

Table 2.
Chemical composition and mechanical properties of chosen products of microalloys steels for forged elements [5-7]

Steel type	Component contents, %								Mechanical properties			
	C	Mn	Si	P	S	Nb+V	Nb	other	R_e MPa	UTS MPa	DVM J	KV J
Metasafe D800[1]	0,15÷0,25	1,3÷2,0	0,1÷0,4	<0,030	<0,015	0,10÷0,20	-	-	-	750÷900	-	-
Metasafe D900[1]	0,25÷0,30	1,3÷2,0	0,1÷0,4	<0,030	<0,015	0,10÷0,20	-	-	-	850÷1000	-	-
Metasafe D1000[1]	0,35÷0,50	1,3÷2,0	0,1÷0,4	<0,030	<0,015	0,10÷0,20	-	-	-	950÷1100	-	-
49MnVS3[2]	0,44÷0,50	0,70÷1,0	≤ 0,50	≤0,035	0,030÷ 0,065	0,08÷0,13 V	-	-	≥450	750÷900	15÷30	-
27MnSiVS6[2]	0,25÷0,30	1,30÷1,6	0,50÷08	≤0,035	0,030÷ 0,050	0,08÷0,13 V	-	(Ti)	≥500	800÷950	40÷60	-
38MnSiVS5[2]	0,35÷0,40	1,20÷1,5	0,50÷08	≤0,035	0,030÷ 0,065	0,08÷0,13 V	<0,050	(Ti)	≥550	820÷1000	20÷45	-
44MnSiVS6[2]	0,42÷0,47	1,30÷1,6	0,50÷08	≤0,035	0,020 ÷0,035	0,10÷0,15 V	-	(Ti)	>600	950÷1100	20÷30	-
HV080SL[3]	0,41÷0,48	0,6÷1,0	0,15÷0, 3	≤0,035	0,020÷ 0,040	0,08÷0,13 V	0,04÷ 0,06	-	>500	>800	-	-
HV090SL[3]	0,48÷0,54	0,8÷1,1	0,20÷03	≤0,035	0,020÷ 0,040	0,08÷0,13 V	0,04÷ 0,06	-	>500	>900	-	-
25GVN[4]	0,25÷0,29	1,3÷1,4	<0,35	≤0,022	0,010÷ 0,030	0,10÷0,15 V	-	0,020N	560-690	700÷900	-	>25
									>740*	>830	-	>45
C-Mn-Ti-V-N[5]	0,25÷0,35	1,5÷1,8	0,4÷0,7	≤0,040	0,030÷ 0,080	0,08÷0,15 V	≤0,02	0,015÷ 0,025N	>620	>820	-	>38

[1] – steels developed in France could be produced with increased contents of S up to 0,045% or < 0,25% Si and ≤ 0,25% Pb to increase machinability; [2] – steels produced by Thyssen company; [3] – steels developed by Fiat Auto and Deltasider Company; [4] – steel worked in Silesian University of Technology; [5] – steels developed in Institute of Ferrum Metallurgy in Gliwice; DVM – energy to break a sample with intersection 10x10 mm with U notch with 3mm depth , * - forging with bainitic structure

Especially usefull for this are steels with micro-additives of B which increase hardenability, and Ti which makes a shield for this element preventing of his connection with N to stable nitride BN. Anyway, when the concentration of Ti is too small to absorbe completely N in this case B is cut out from his effecting on hardenability of the steel and at the same moment not enough volume of TiN will not be a barrier for growing austenite grains in high temperature during heating up the charge.

2. Experimental procedure

The goal of this experiment is a structure and mechanical properties of forged elements using thermo-mechanical treatment, made from C-Mn steel with micro-additions of Ti, V, B and N (table 3) melted using: after furnace treatment of metal bath, and continuous casting ingots with intersection 100×100 mm. Ingots after their solidifications were adjustly rolled for bars with a dimention about 36 mm, and after using the same conditions for bars with dimention of 17 mm. The range of adjusted rolling temperature were taken basing on calculation of solubility in the austenite microadditions added to the steel. To make this kind of calculation authors used a kinetic equation (1) and data from table 1.

It was found that micro-addition of Ti introduced to the steel type A with concentration of 0,005% is completely dissolved in austenite in temperature about 1250 °C (fig. 2) and during cooling is not absorbing whole N from the steel to TiN. In this case excess of N is creating nitrides of AlN and BN during cooling the steel (fig. 3). Boron is fixed to stable nitride BN and practically is not increasing the hardenability of the steel. On the other hand nitride from steels B and C type is completely fixed in TiN. Total dissolving of that phase in the austenite is possible only in temperatures of 1400 and 1450 °C, respectively for steel type B and C. The rest of introduced to those steels amount of Ti is fixing itself to carbonitrides Ti(C,N) and carbides TiC. However, Al stays in dissolved state and B is fixing to $M_{23}(C,B)_6$ during cooling of the steel which could be dissolved in the solid solution in temperatures little higher then temperature A_{c3} of the steel.

These data related with kinetics of precipitation processes of MX phases in austenite, were used to determine conditions of adjusting rolling of charge for forging in a shape of bars within the temperature range from 1200 up to 900 ºC [8, 9].

Table 3.
Chemical composition of tested steels

Steel	Chemical composition, %												
	C	Mn	Si	P	S	Cr	Ni	Mo	Ti	V	B	Cu	Al_c	N
A	0,21	1,02	0,25	0,018	0,005	0,18	0,07	0,03	0,005	0,008	0,002	0,11	0,024	0,009
B	0,21	1,03	0,23	0,018	0,013	0,14	0,07	0,03	0,032	-	0,002	0,14	0,024	0,008
C	0,32	1,17	0,25	0,015	0,007	0,22	0,07	0,02	0,035	0,026	0,003	0,23	0,025	0,010

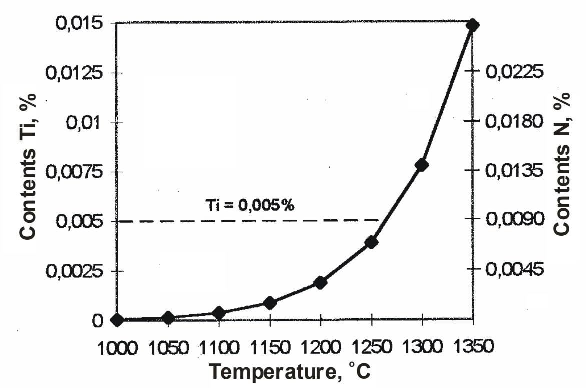

Fig. 2. Solubility curve of TiN nitride in austenite of A type steel in function of temperature

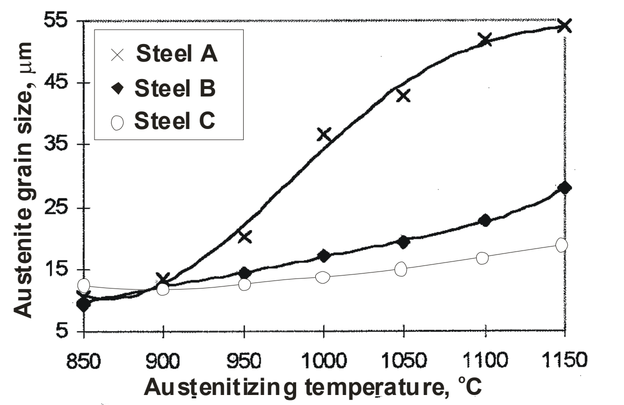

Fig. 4. Influence of the austenitizing temperature on the grain size of austenite

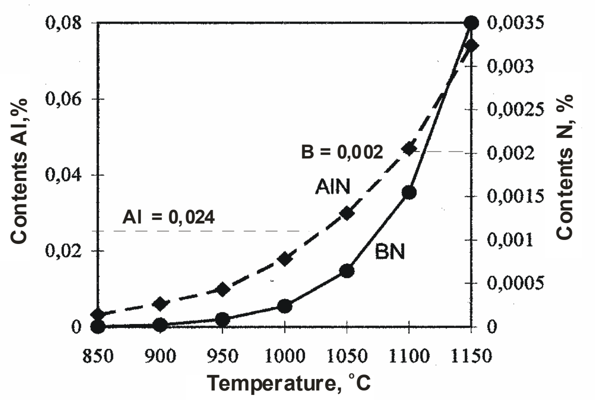

Fig. 3. Solubility curves of AlN and BN nitrides in austenite of A type steel in function of temperature

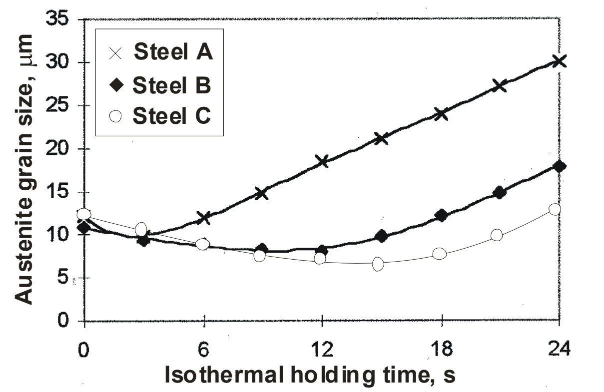

Fig. 5. Influence of the isothermal holding time of the specimens after completing hot-working at a temperature of 900°C before water quenching on the grain size of austenite

Charge heating temperature for forging was calculated basing on primary austenite grain size index of the samples quenched from programmed increasing austenitizing temperature (fig. 4).

As it could be seen steel A type with small concentration of Ti and not high volume of TiN phase have distinct grow of austenite grains just after crossing 900 ºC temperature.

Table 4.
Results of the mechanical properties and hardenability of the investigated steels

Steel	Treatment type		$R_{p0,2}$ MPa	UTS MPa	A %	Z %	KV J	HB	D_I mm
	Treatment conditions	Tempering temperature, °C							
A	900°C/3s/water	600	695	770	22	69	186	230	28
B	900°C/12s/water	600	700	786	24	75	197	220	55
C	900°C/16s/water	500	1021	1115	13	50	72*	275	96
		600	852	932	18	65	108*	250	

* - specimen breaking energy tested at temperature of – 20 °C

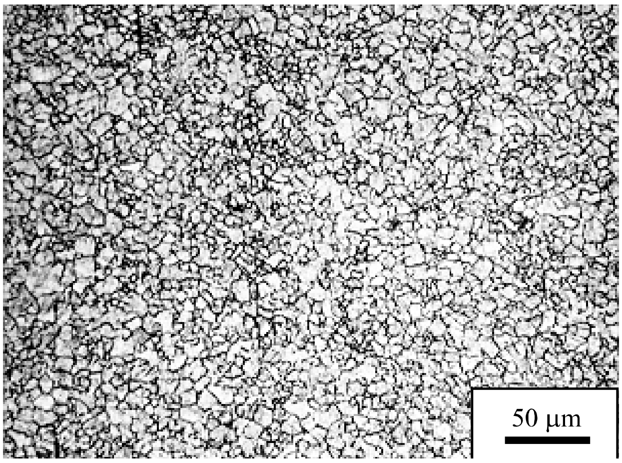

Fig. 6. Fine-grained structure of the statically recrystallized austenite; finishing forging temperature 900°C; isothermal holding time 12s; steel B

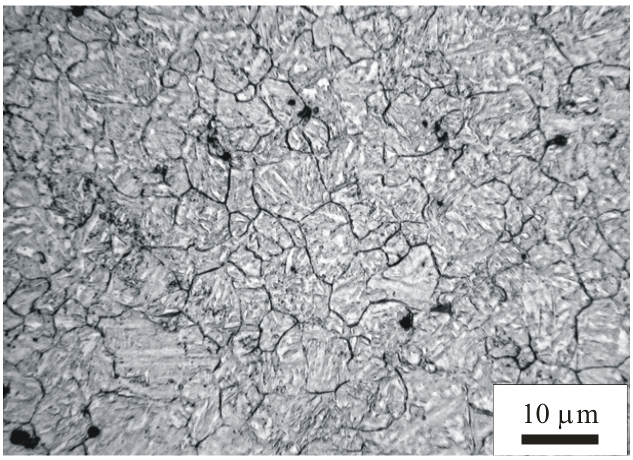

Fig. 7. Fine-grained structure of the statically recrystallized austenite; finishing forging temperature 900°C; isothermal holding time 16s; steel C

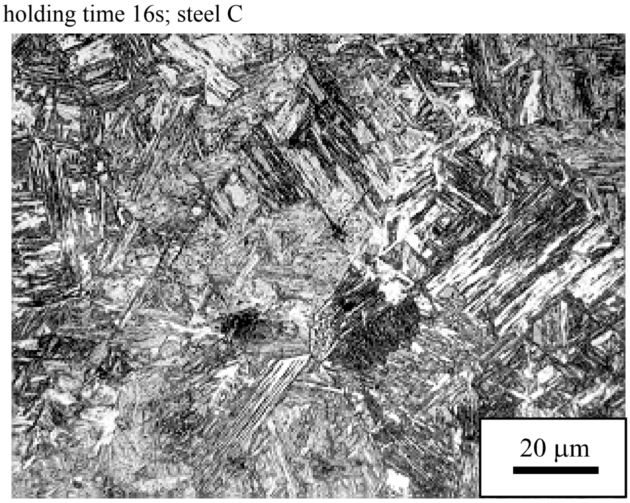

Fig. 8. Martensitic-bainitic structure of steel B water quenched from a finishing hot working temperature of 900°C after holding for 12s

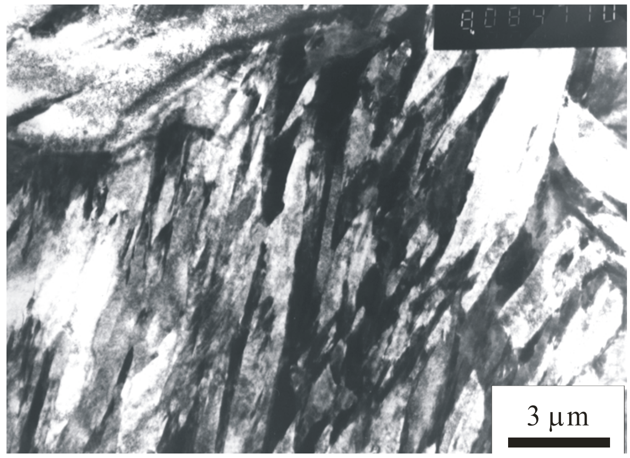

Fig. 9. Lath martensite structure of steel C quenched from a finishing hot working temperature of 900°C after holding for 16s

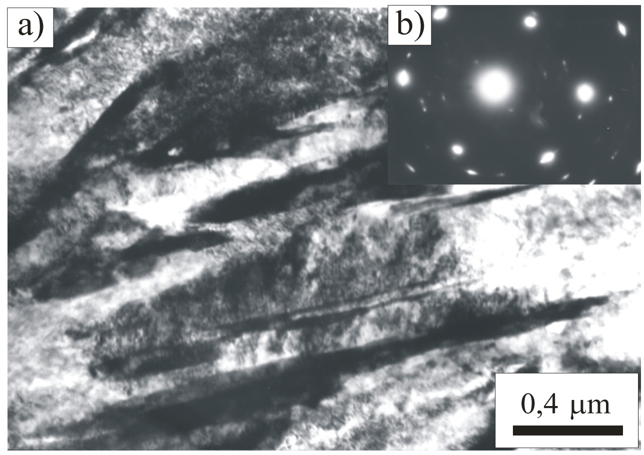

Fig. 10. Lath martensite structure with the dispersive Fe_3C precipitations (steel C): a – light field, b – diffraction pattern to Fig. 10a [111]Feα and [012]Fe_3C

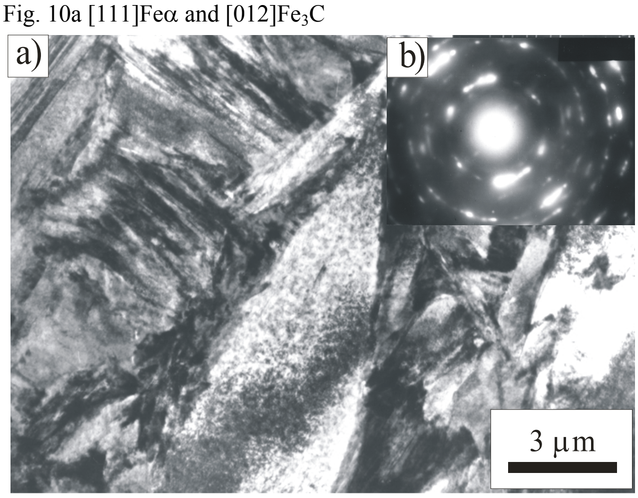

Fig. 11. Dispersive $M_{23}(C,B)_6$ precipitations at the grain boundaries of the primary austenite (steel C); a – light field, b – diffraction pattern to Fig. 11a [111]Feα and [001]$M_{23}(C,B)_6$

Whereas growth of grains of the γ phase of the steel type B proceed softly up to 1100° C temperatures, and in steel type C even higher than 1150°C. Basing on a data which were shown charge heating temperatures were adjusted for steel A type - 950°C, and for steel B and C respectively 1000 and 1150°C.

Time needed for occuring static recrystallization of γ phase after finishing plastic deformation in the temperature of end forging was determined basing on a primary austenite grain size of upset forging samples with $\varepsilon = 0{,}25$ in a temperature of 900°C with deformation speed of 14 s^{-1}, hold before quenching in the water for the time from 0 up to 24 s (fig. 5). As it is shown in that figure steel A type obtains the finest austenite grain size after holding samples in this temperature for 3 s, whereas steels B and C type for 12 s and 16 s respectively.

Thermo-mechanical treatment was realized thanks to open die forging of experimental segments with 17 mm diameter and 150 mm length in the temperature range of 950 to 900°C and 1000 to 900°C - respectively from steel A and B type to rods with intersection 12 x 12 mm. Before quenching in water these rods were hold at a temperature of end forging for 3 and 12s. Quenched rods were tempered at a temperature of 600°C for 1h. Whereas rod segments from steel type C with intersection 24 x 24 mm were forged in a range of temperature from 1150 up to 900 °C to the shape of rods of intersection 12 x 12 mm. Before quenching in water the rods were hold in a temperature of end forging for 16s, and after were tempered in temperature of 500 and 600°C for 1h.

Investigations showed that examinated steels after thermo-mechanical treatment and after quenching have a fine-grain structure of primary austenite (fig. 6, 7) with a grain size about 10, 5 and 8 and martensitic-bainitic structure (fig. 8) and hardness 42, 44 and 49 HRC – respectively for a steel A, B and C type.

Hardness of the steels is decreasing after high tempering from 220 to 250 HB, which not create difficulties, during mechanical treatment of forged elements.

Examinations of the structure of thin foils made in TEM (transmission electron microscope) showed that steel C type (quenched from a temperature of end forging – 900 °C and hold before that in this temperature for 16s) have lath martensite structure (fig. 9). Inside of martensite laths it was approved a presence of dispersive particles of cementite (fig. 10), whereas at the boundaries of primary austenite $M_{23}(C,B)_6$ type dispersive particles were found (fig. 11), which have occure in a steel during self tempering process.

Particular attention should be focused on high mechanical properties of the steels in high tempered state and especially on their crack resistance also in low temperatures (table 4). In this table also shown ideal diameter D_I, which characterizing their hardenability. The ideal diameter was calculated due to procedure described in ASTM A255-89 standard. Micro-additive of boron introduced to the steel A type is totally fixed in nitride BN and because of this is not increasing hardenability of the steel. On the contrary in steels B and C type micro-additive of boron dissolved in solid solution is affecting on hardenability. It was noticed during calculated ideal diameter D_I, which for the steel type A is equal 28 mm.

At the same time calculated ideal diameter for steels type B and C taking into account influence of micro-additive of boron D_{IB} is equal respectively - 55 and 96 mm. This sugests that steel A type is useful for forging relatively small intersections and the steels B and C type for a big one [10].

3. Conclusions

Investigations carried out shown full usability of micro-alloyed steels for producing forged machine parts, with high strength and good cracking resistance, using energy-saving thermo-mechanical treatment method. This thermo-mechanical treatment, allows to obtain fine-grained austenite structure during hot plastic deformation, and gives us forged elements obtaining yield point $R_{p0,2}$ over 690 MPa, UTS over 770 MPa, hardness 220 up to 250 HB and breaking energy KV over 180J. Those properties could be reach after holding forged elements in temperature of end plastic processing, for the time needed to create at least 50% recrystallized austenite, after they have to be quenched from this temperature and finally highly tempered.

Designing of forging technology from micro-alloys steels needs accommodation of charge heating conditions to the kinetics of precipitation processes of MX phases in austenite, without growing grains of γ phase. Thanks to investigations of influence of austenitizing temperature on a primary austenite grain size, and a kinetics of disolving interstitial MX phases in austenite it was found that charge heating temperature for forging elements made from steel A type should not exceed 950 °C, and for elements made from steels B and C type could have even 1150 °C.

This shows that a charge made from steels B and C type could be heated to temperature considerably higher than A_{c3} for that steel, with keeping fine-grained structure. This is increasing the live time of dies. Introducing micro-additives of boron to the steel (which increase the hardenability) made from fine-grained steel structure needs a Ti shield with quantity indispensable to fix all nitrogen to BN. When the concentration of Ti is too small to consume whole nitrogen from the steel in that moment boron is turn off from effecting on hardenability of investigated steels. Also small amount of TiN nitrides is not protecting steel from austenite grains grow, in a high temperature used for charge heating.

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