EFFECT OF PRIOR WORKING ON THE STRUCTURE AND DEFORMATION CAPACITY OF Mn–29.5% Al–0.5% C ALLOY

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Alloys based on the system Mn–Al–C relate to promising magnetically hard materials which in magnetic properties in an anisotropic state are close to barium ferrites, and in magnetic energy referred to density, they markedly surpass widely known alloys of the YuNDK type [1]. The high magnetic properties of the system Mn–Al–C are obtained as a result of plastic deformation [1-3]. Metastable τ-phase, governing the ferromagnetic properties of these alloys, is stable up to 700-750°C, and therefore deformation treatment of them is not carried out at higher temperatures. It should be noted that alloys of the Mn–Al–C system have high brittleness, low ductility, and poor workability. For example, in order to obtain magnets of these alloys by extrusion at 700°C a stress of more than 800 N/mm² is required [2], which leads to rapid wear of the die tool. A study was made in [4, 5] of alloys containing lamellar τ-phase of the martensitic type obtained after air cooling. The morphology of the τ-phase may be altered by using special working methods, e.g., extrusion [3] or prior plastic deformation as a result of which it is possible to expect a considerable improvement in the working capacity of the alloy. It is of interest to study prior deformation of an alloy of the Mn–Al–C system in the temperature region for existence of high-temperature ε-phase followed by cooling.

The aim of the present work is to study the effect of prior plastic deformation on the working capacity of alloy Mn–29.5%Al–0.5%C [2–3].

Studies were carried out on specimens 12 mm in diameter made with casting according to a dispensable model. Melting and casting were carried out in a protective atmosphere. Ingots were homogenized at 1150°C for 2 h and subsequently air checked according to saturation magnetization $4\pi I_S$ is in a field with intensity H = 144 kA/m. It was established that the time for τ-phase stability at 700°C, $\tau_{sta} > 60$ min, and $4\pi I_S = 0.57$ T. At 750°C, $\tau_{sta} = 30$ min.

Results of metallographic and qualitative x-ray analyses showed that after annealing at 700°C for 60 min and 750°C for 30 min only τ-phase is observed in specimens.

Homogenized specimens 12 mm in diameter and 18 mm high were extruded with a degree of deformation of 30 and 80% at 700°C and at 950°C, i.e., the temperature corresponding to the single-phase region of the phase composition diagram [6]. Extrusion was carried out under isothermal conditions by different regimes (Table 1). Graphite was used as a lubricant.

The working capacity of the alloy in the original homogenized conditions was determined by free upsetting in flat hammers in a 1231 U 10 machine under isothermal conditions in the temperature range 650-750°C with deformation rates $v_d = 5\cdot10^{-3}$ and 10^{-2} sec^{-1}. Upsetting of specimens made from extruded bars was carried out at 500-750°C and $v_d = 10^{-1}$-10^{-3} sec^{-1}.

Metallographic studies were carried out in a Neophot-2 light microscope and in a Tesla BS-540 transmission electron microscope, and qualitative x-ray analysis was carried out in a DRON-2 diffractometer in chromium radiation using a vanadium filter.

The structure of specimens in the original condition is packets of τ-phase platelets forming as a result of shear $\varepsilon\to\tau$-transformation [4] and located within the bounds of previous ε-phase grains. The structure of the τ-phase is characterized by thin uniformly orientated platelets parallel to the close-packing planes of the HCP-lattice of ε-phase (Fig. 1a). In addition, there are transformation and dislocation twins in the structure whose density is 10^{-7}-10^{-9} cm^{-2}.

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TABLE 1

Extrusion regime	t, °C	ε, %	σ_{max}, N/mm^2	Phase composition*
1	700	80	840	τ
2	700	30	300	τ
3	950	80	320	ε
4	950	30	100	ε

*At the extrusion temperature.
Note. The deformation rate in all regimes v_d = 2 mm/min.
Notation: σ_{max} is maximum specific load.

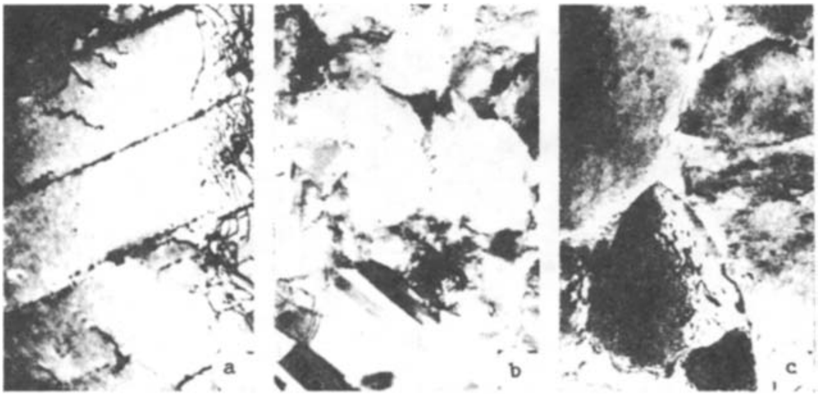

Fig. 1. Microstructure of 29.5% Mn–0.5% C alloy. ×12,000; a) in the original condition; b, c) after extrusion with ε = 80% at 700 and 950°C respectively.

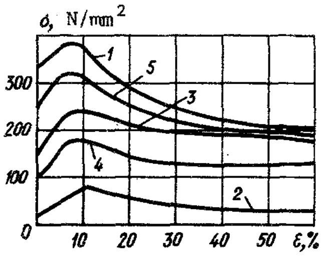

Fig. 2. Curves of flow stress (σ)—degree of deformation (ε) with upsetting Mn–29.5% Al–0.5% C alloy at 700°C and a deformation rate $v_d = 10^{-2}$ sec^{-1}; 1) after homogenizing; 2) after extrusion at 700°C, ε = 80%; 3) 700°C, ε = 30%; 4) 950°C, ε = 80%; 5) 950°C, ε = 30%.

As a result of qualitative x-ray and magnetic analyses it was established that after warm and hot working the structure of specimens only consists of τ-phase.

After extrusion by regime 1 τ-phase has an ultrafine grained (d_3 = 0.5-2 μm) structure (Fig. 1b). The dislocation density for almost half of the grains is low (10^6-10^7 cm^{-2}). No τ-phase platelets are observed, and in a number of grains twins are detected.

The structure of the alloy after extrusion by regime 2 occupies an intermediate position between the original structure and that obtained by deformation with regime 1. Mainly τ-phase platelets are observed located in the form of packets, and areas with a granular structure occupy not more than 10% of the microsection area.

After extrusion at a temperature of the ε-region (regime 3) and air cooling τ-phase is observed in specimens having a granular structure (d_3 = 10-15 μm) and there are almost no τ-phase platelets in the alloy (Fig. 1c). The dislocation density in grains is about 10^7-10^8 cm^{-2}.

TABLE 2

Extrusion regime	t_{up}, °C	v_d, sec^{-1}	σ_{max}	$\sigma_{s.s}$
			N/mm^2	
1	500	10^{-1}	870	630
		10^{-2}	610	510
		10^{-3}	390	350
	550	10^{-1}	563	399
		10^{-2}	295	278
		10^{-3}	180	173
	600	10^{-1}	347	246
		10^{-2}	122	120
		10^{-3}	82	82
	650	10^{-1}	253	174
		$5\cdot10^{-2}$	200	125
		10^{-2}	102	90
		$5\cdot10^{-3}$	62	62
		10^{-3}	31,5	31,5
	700	10^{-1}	152	109
		$5\cdot10^{-2}$	121	80
		10^{-2}	69	58
		$5\cdot10^{-3}$	45	40
		10^{-3}	21	20,5
	750	10^{-1}	114	84
		$5\cdot10^{-2}$	81	72
		10^{-2}	60	51
		$5\cdot10^{-3}$	36	35
		10^{-3}	18,9	19,1
2	700	10^{-2}	250	180
		$5\cdot10^{-3}$	155	100
3	700	10^{-2}	180	135
		$5\cdot10^{-3}$	165	125
4	700	10^{-2}	435	280
		$5\cdot10^{-3}$	348	230

Notation: σ_{max} is maximum stress in steady stage.

The structure of τ-phase after extrusion by regime 4 occupies an intermediate position between the structures in the original conditions and after treatment by regime 3. It should be noted that the length of τ-phase platelets is less here than the original, which is probably connected with crushing ε-phase during working.

We consider the test results for alloy in different conditions with upsetting (Fig. 2). Of greatest interest are properties at 700°C since at lower temperatures the alloy is difficult to work, and at temperatures above 700°C τ-phase stability decreases.

It was established that the maximum flow stress (σ_{max}) depends on the alloy conditions. For example, after homogenizing σ_{max} = 390 N/mm², which agrees with data in [7], and after extrusion with ε = 80% at 700°C, σ_{max} = 58 N/mm².

The relationship σ = f(ε) in the homogenized condition is characterized by a high flow stress level at the start of deformation, and then rapid loss of strength. Curve 1 has a maximum corresponding to ε = 8%. From the nature of this curve it is possible to suggest that during working of the alloy there is dynamic recrystallization; this is confirmed by data in [3].

Prior working by extrusion promotes a reduction in flow stress compared with homogenizing, and the nature and degree of loss of strength is governed by prior working conditions. For example, deformation in the ε-region with a degree of 30% leads to a reduction in σ_{max} by about 20% and to a reduction in the degree of loss of strength compared with homogenizing. Alloy extrusion at 700°C and ε = 30% causes a reduction in σ_{max} by 35%, which even more reduces the loss of strength effect. The maximum degree of reduction in flow stress is achieved after alloy deformation with ε = 80%. As a result of upsetting specimens worked by regime 3 it was established that prior treatment promotes a reduction by more than a factor of two of the maximum flow stress and in the steady stage by more than a factor of 1.5. The degree of loss of strength also decreases compared with the cast condition.

The lower flow stress occurs for specimens extruded by regime 1. Curve 5 has almost no strengthening stage. Flow stress in the steady stage is lower by a factor of more than six than the flow stress for the alloy in the homogenized condition.

Results of upsetting tests for specimens extruded by regimes 1-4 (Table 1) over a wide temperature-rate range are given in Table 2. It can be seen that prior working over the whole test temperature-rate range promotes a reduction in σ_{max} and flow stress in the steady stage. The lowest flow stress occurs for specimens after extrusion by regime 1. Alloy after this treatment exhibits high deformability in the temperature range 550-750°C and at deformation rates of 10^{-1}-10^{-3} sec^{-1}. In determing the rate sensitivity parameter for flow stresses $m = \frac{d \ln \sigma}{d \ln \varepsilon}$ (where σ is flow stress) it was established that in the range 650-750°C and $v_d = 10^{-3}$-10^{-2} sec^{-1} parameter m = 0.4-0.5, which is typical for materials in the superelasticity condition [8]. This conclusion agrees with data for structural studies. As was noted above, the structure of the alloy treated by regime 1 is characterized by the presence of ultrafine τ-phase grains.

The dependence of flow stresses on temperature and deformation rate for specimens after extrusion by other regimes is similar, but notably weaker. An increase in deformation rate at 700°C (regime 2) from $5 \cdot 10^{-3}$ sec^{-1} to 10^{-2} sec^{-1} leads to an increase flow stress in the steady stage from 100 to 180 N/mm^2.

Thus, the results of studies carried out have shown that prior working causes a change in τ-phase morphology. Occurrence of fine-grained τ-phase after extrusion of cast alloy at 700°C with ε = 80% is apparently connected with processes of dynamic recrystallization which were observed in these alloys in [3] and confirmed by the results of analyzing the nature of σ = f(ε) curves. Formation of individual τ-phase grains after deformation of alloy with ε = 30% at 700°C is also evidently a result of alloy recrystallization processes. According to the data obtained it is possible to conclude that working in the temperature region for existence of τ-phase promotes creation of an ultrafine grained structure for the alloy.

Working of alloy in the single-phase ε-region of the phase composition diagram causes ε-phase grain refinement. In alloy quenched from the homogenizing temperature the grain size of ε-phase is 500-1000 μm. Shear ε→τ-transformation [4] leads to the situation that in coarse ε-phase grains long rough τ-phase platlets form. After hot plastic deformation coarse ε-phase grains are refined, and therefore τ-phase obtained in refined ε-phase grains is more dispersed.

In particular, occurrence of τ-phase of granular shape after deformation with ε = 80% at 950°C creates a change in the mechanism of ε→τ-transformation. It may be suggested that shear transformation changes into diffusion transformation, but this question requires special studies.

CONCLUSIONS

1. Alloy Mn—29.5%Al—0.5%C in the homogenized condition at temperatures below 650°C exhibits low working capacity, i.e., high flow stress and brittleness, which is due to presence in the alloy of lamellar τ-phase. Upsetting of the alloy is possible only in the temperature range 650-750°C and at deformation rates $5 \cdot 10^{-3}$-10^{-2} sec^{-1}.
2. As a result of extrusion at 700°C recrystallized fine τ-phase grains are formed with a size of 0.5-2 μm. With extrusion in the single-phase ε-region (950°C) there is ε-phase grain refinement, and then during subsequent cooling fine-grained τ-phase forms. Deformation by extrusion with ε = 80% promotes total transformation of lamellar τ-phase into granular phase which causes an improvement in the working capacity of alloy Mn—29.5%Al—0.5%C.
3. In alloy Mn—29.5%Al—0.5%C, containing τ-phase with a grain size of 0.5-2 μm the superplasticity effect is observed which is more clearly developed with t = 650-750°C and $v_d = 10^{-2}$-10^{-3} sec^{-1}.

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RETURNING TO PUBLISHED ARTICLES

EFFECT OF NITROGEN ON THE PHASE COMPOSITION AND MECHANICAL PROPERTIES OF CORROSION-RESISTANT AUSTENITIC-FERRITIC STEELS

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An article by G. G. Kolchin, N. V. Korolev, and B. S. Ermakov "Effect of nitrogen on the phase composition, physical, and mechanical properties of corrosion-resistant austenitic-ferritic steels" was published in the journal *Metallovedenie i Termicheskaya Obrabotka Metallov* [1]. The main position and conclusions of this article cause doubts and contradict already existing results of studying the effect of nitrogen on the phase composition and mechanical properties of chromium-nickel-molybdenum and chromium-nickel-manganese corrosion-resistant steels.

The authors of [1] studied steels of three compositions which differed in manganese, molybdenum, and nitrogen content (Table 1). It is well known that the solubility of nitrogen in steel depends mainly on Cr, Mn, and Ni content. Addition of Cr and Mn promotes an increase in the solubility of nitrogen in iron, but Ni promotes a reduction of it. The authors of [1] confirm that with a constant content of Cr (21.5%), Ni (11.5%) and a reduced content of Mn from 4.8 to 0.9% in steel 3 it is possible to introduce 0.36% N. We are apparently talking about the solubility of nitrogen in solid solution since Al and V are elements with which nitrogen could form nitrides or carbonitrides, found completely in solid solution both in steel 1 (without nitrogen) and in steel 3 (0.36% N) (Table 2). It is logical to assume that in the test steel with additions of nitrogen apart from γ- and α-phase, secondary austenite (γ'), σ-phase, carbides $M_{23}C_6$, noted by the authors, aluminum nitride AlN and nitrides or vanadium carbonitrides VCN are present which would precipitate during cooling after forging. However, in [1] there is no mention of the presence or possibility of forming these phases. To suggest that during exposure at the heating temperature for hardening 850-1100°C there was dissolution of nitrides and carbonitrides is not substantiated since the dissolution temperature for aluminum nitrides is 1100-1350°C, and dissolution in austenite of vanadium nitrides and carbonitrides with a content of 0.29% V proceeds at 1200°C [2]. Thus, exposure even at 1100°C could not lead to total dssolution of aluminum and vanadium nitrides and carbonitrides if they total dissolution of aluminum and vanadium nitrides and carbonitrides if they occurred in these steels.

A study of nitrogen as a alloying element normally starts with determination of the limit of its solubility in steel of a given composition. A review of procedures for calculating nitrogen content corresponding to its solubility in chromium-nickel-manganese and other steels has been given in [3]. In our developments of cast chromium-manganese steels alloyed with nitrogen [4-6] we used the Langenberg method [7] by which the nitrogen activity factor f_N is determined as a function of the concentration of different alloying elements entering into solution at 1600°C and a nitrogen pressure of 0.1 MPa. Then an estimate is made of the maximum nitrogen content in the steel from a graphical dependence of nitrogen solubility on the factor of its activity. Practice has shown that this method guarantees with observation of all of the normal production requirements preparation of dense castings without gas poro-

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