Metal Science and Heat Treatment, Vol. 55, Nos. 9 – 10, January, 2014 (Russian Original Nos. 9 – 10, September – October, 2013)

UDC 621.785.3:669.715.017:620.18:622.24.05

STRUCTURAL STABILITY AND VARIATION OF PROPERTIES OF ALUMINUM ALLOYS D16 AND 1953 IN PRODUCTION AND OPERATION OF DRILL PIPES

S. Yu. Kondrat'ev,[1] O. G. Zotov,[1] and O. V. Shvetsov[2]

Translated from *Metallovedenie i Termicheskaya Obrabotka Metallov*, No. 10, pp. 15 – 21, October, 2013.

The effect of the temperature and time parameters of production and service of drill pipes from aluminum alloys D16T and 1953T1 on their structure and mechanical properties is studied. The structural changes in the alloys due to production and operation heatings are determined.

Key words: aluminum alloys, microstructure, mechanical properties, aging, drill pipes.

INTRODUCTION

Today, aluminum alloys compete with steels successfully in the production of equipment for the oil and gas industry due to their low specific weight, high mechanical properties and corrosion resistance [1 – 4]. For example, drill pipes are fabricated from deformable alloys D16T and 1953T1 a considerable advantage of which is their high specific strength. This characteristic depends substantially on the production process and on the conditions of operation of the equipment. Shrink fitting of an interlock heats the metal to 250°C, while drill pipes in service are heated to 100 – 200°C [3, 4]. The operating capacity of aluminum alloys under such conditions is determined to a high degree by their heat resistance. However, we should state that the relevant literature contains no systematized experimental data on the effect of the production process and operating conditions on the properties of aluminum drill pipes.

The aim of the present work was to study the structure and mechanical properties of drill pipes from aluminum alloys D16T and 1953T1 under successive action of production (hot fitting of interlocks) and operational heatings.

METHODS OF STUDY

Preforms for the study were cut in the longitudinal direction from deformed semiproducts, i.e., pipes with external diameter 147 mm and wall thickness 13 mm produced from aluminum alloys D16T and 1953T1 of the Al – Cu – Mg and Al – Zn – Mg systems respectively. As-delivered, the pipes were heat treated in a standard mode. For D16T the treatment consisted of water quenching from 500°C and natural aging for 4 days; for 1953T1 the treatment consisted of water quenching from 480°C and artificial aging at 125°C for 24 h.

The homogeneity of the chemical composition of the materials was estimated by microscopic x-ray spectrum analysis using a "Camebax" device. Studies of various pipe regions showed the absence of segregation of the alloying elements. The content of the main chemical elements in the alloys matched the standardized values (Table 1).

We prepared standard fivefold cylindrical specimens 6 mm diameter for the tensile mechanical tests and cylindrical specimens 10 mm in diameter and 20 mm high for the x-ray diffraction and metallographic studies and for determining the microhardness.

[1] St. Petersburg State Polytechnic University, St. Petersburg, Russia (e-mail: petroprom2013@yandex.ru).

[2] St. Petersburg Branch of the "Akvatik" Company, Moscow, Russia.

TABLE 1. Actual Chemical Composition of Studied Alloys

Grade of alloy	Content of main chemical elements, wt.%									
	Al	Mg	Zn	Mn	Cu	Zr	Cr	Ti	Fe	Si
D16T	Base	1.62	0.30	0.53	4.54	–	–	0.08	0.45	0.48
1953T1	Base	2.60	5.70	0.17	0.45	0.02	0.19	0.05	0.10	0.05

The specimens were subjected to a single short-term hold for 3 – 10 min at 200 or 250°C to imitate the heating of drill pipes due to hot fitting of interlocks and then to a long hold at 150°C for 500 or 1000 h, which matched the operating conditions. The specimens were heated in "SNOL-1.6.2.5.1/11-I2" chamber furnaces. The temperature was controlled with the help of a thermocouple calked into the center of the control specimen. The accuracy of the temperature control was ± 2.5°C. The specimens were charged into the furnace preliminarily heated to the specified temperature, held for the specified time, and then cooled in air. After the heat treatment we determined the mechanical properties and studied the structure of the alloys.

Metallographic analysis of longitudinal and transverse laps was performed with the help of a "Reichert-Jung MeF3A" light microscope at magnifications × 100 – × 500 using a software for automatic quantitative analysis of images in accordance with the ASTM E 1245-03 standard. The metallographic specimens were prepared using a "Buehler" facility in accordance with the ASTM E 3-95 standard.

Phase analysis was performed with the help of a universal "Bruker D8 Advance" x-ray diffractometer under radiation of a tube with a copper anode. The diffractograms were obtained at a voltage of 4 kV, a current of 40 mA, a step of the scintillation detector of 0.01, and an interval of Bragg diffraction angles $2\theta = [15 – 130]°$ for dominant orientations (111) and (002).

The mechanical properties were determined in static tests for uniaxial tension at room temperature according to the GOST 1497–84 standard. The tests were performed with the help of a "Schenck" machine with maximum load of 200 kN at a loading rate of 1 mm/min.

The microhardness was determined by the Vickers method using a "Reichert-Jung Micro-Duromat 4000E" device in accordance with ASTM E 92.

RESULTS AND DISCUSSION

The results of the determination of the mechanical properties of the metal of the pipes in the state as-delivered are presented in Table 2. It can be seen that the properties of the alloys virtually match the standardized values. Note that the initial strength of alloy 1953T1 is higher and the ductility is lower than those of D16T.

The variation of the strength and ductility of the alloys after process heating has been studied in detail in our work [5]. We established that process heating within 200 – 250°C for up to 10 min in hot fitting of an interlock affects substantially the structure and mechanical properties of the metal of drill pipes from aluminum alloys D16T and 1953T1. Alloy D16T is much more resistant to process softening than alloy 1953T1.

The analysis of structures performed in [5] has shown that the variation of the properties of a drill pipe under process heating is caused by diffusion processes that occur differently in the two alloys. As a result of the contradictory effect of the "aging retrogression" and subsequent precipitation hardening the properties of D16T remain virtually invariable after the heating as compared to the state as-delivered. The decomposition of the solid solution and the coarsening of the particles of intermetallic phases in heating of alloy 1953 in the studied range of holds causes considerable softening.

TABLE 2. Actual and Standardized Mechanical Properties of the Studied Alloys As-Delivered

Grade of alloy	Material	σ_r, MPa	$\sigma_{0.2}$, MPa	δ, %	ψ, %
D16T	Actual	520	340	15	14
	GOST 4784–97	≥ 300	≥ 470	≤ 19	–
1953T1	Actual	570	520	12	21
	TU 1-2-592–2003	≥ 520	≥ 470	≤ 6.5	–

Basing ourselves on the experimental data on the effect of process heatings on the structure and properties of alloys D16T and 1953T1 we studied the action of subsequent operational heating for specimens treated preliminarily in the most typical modes, namely, heating to 200 and 250°C and holding for 6 and 10 min.

Figure 1 presents the variation of the mechanical properties of alloy D16T after process heating in the chosen modes and subsequent operational heating at 150°C with a hold of 500 and 1000 H. It can be seen that for D16T the 50-h operational hold at 150°C does not affect substantially the ultimate rupture strength σ_r and the 100-h hold decreases it inconsiderably (by 30 – 90 MPa with respect to the state of the material after the process heating depending on its mode and by 80 – 100 MPa with respect to the state as-delivered). The conventional yield strength $\sigma_{0.2}$ of the alloy after the 500-h hold increases inconsiderably (by 20 – 17 MPa) and after the 100-h hold falls to the level corresponding to the state after the process heating. The ductility characteristics of the alloy, i.e., the elongation δ and the contraction ψ, vary monotonically during the hold at 150°C. When the duration of the hold exceeds 500 h, δ and ψ decrease somewhat with respect to the state as-delivered.

It is important that increase in the temperature of the preliminary process heating from 200 to 250°C and in the hold time from 6 to 10 min virtually does not affect the kind and intensity of the variation of the mechanical properties of alloy D16T in the subsequent long-term operation at 150°C.

A 500-h operational hold of alloy 1953T1 at 150°C causes marked and considerable lowering of the strength characteristics, i.e., σ_r decreases by 70 – 110 MPa and $\sigma_{0.2}$ decreases by 20 – 170 MPa with respect to the state of the material after process heating (depending on its conditions) and by 140 – 160 MPa and 180 – 200 MPa, respectively, with respect to the state as delivered (Fig. 2). When the hold time is increased to 1000 h, the alloy softens too, but the intensity of the softening is much lower. The ductility charac-

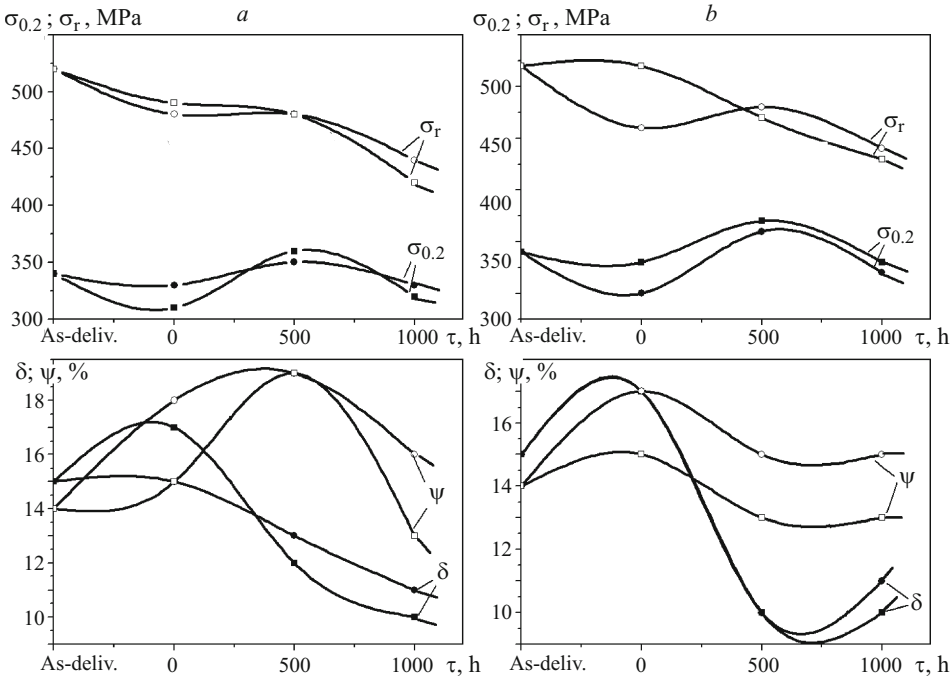

Fig. 1. Effect of the duration of the hold at 150°C on mechanical properties of alloy D16T after preliminary heating (point "0") to 200°C (*a*) or 250°C (*b*) with a hold of 6 min (○, ●) or 10 min (□, ■).

teristics of alloy 1953T1 increase monotonically during holding at 150°C. After a 1000-h hold δ increases by 1 – 3% and ψ increases by 20 – 23% relative to the state of the material after the process heating and by 4 – 7% and 24 – 29%, respectively, relative to the state as-delivered. Note that the growth of the values of contraction is an order of magnitude higher than the growth of the values of the elongation. This indicates decrease in the strain hardening factor of the alloy.

In contrast to alloy D16T, increase in the temperature of process heating from 200 to 250°C and in the hold time from 6 to 10 min promotes substantially higher softening or alloy 1953T1 in subsequent operational heating, especially due to the hold of up to 500 h.

The results of the determination of the mechanical properties show that the heat resistance of D16T is substantially higher than that of 1953T1 both due to the process heating and due to the operational heating of drill pipes. The strength and ductility characteristics of alloy D16T are steady enough under these conditions and change inconsiderably with respect to the state as-delivered (Table 2, Figs. 1 and 2). The

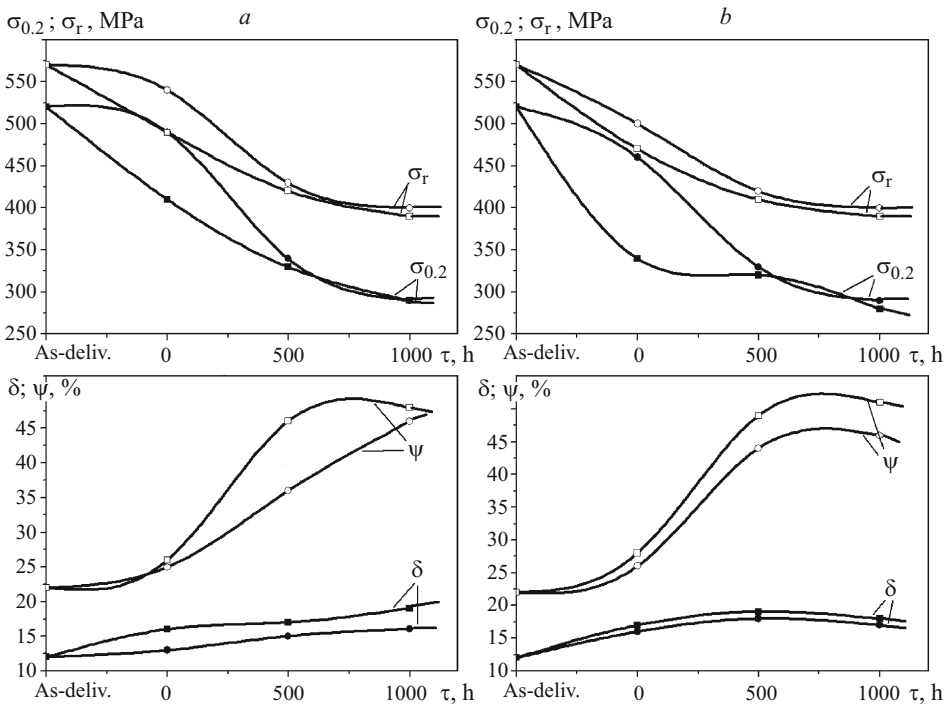

Fig. 2. Effect of the duration of the hold at 150°C on the mechanical properties of alloy 1953T1 after preliminary heating (point "0") to 200°C (*a*) and 250°C (*b*) with a hold of 6 min (○, ●) or 10 min (□, ■).

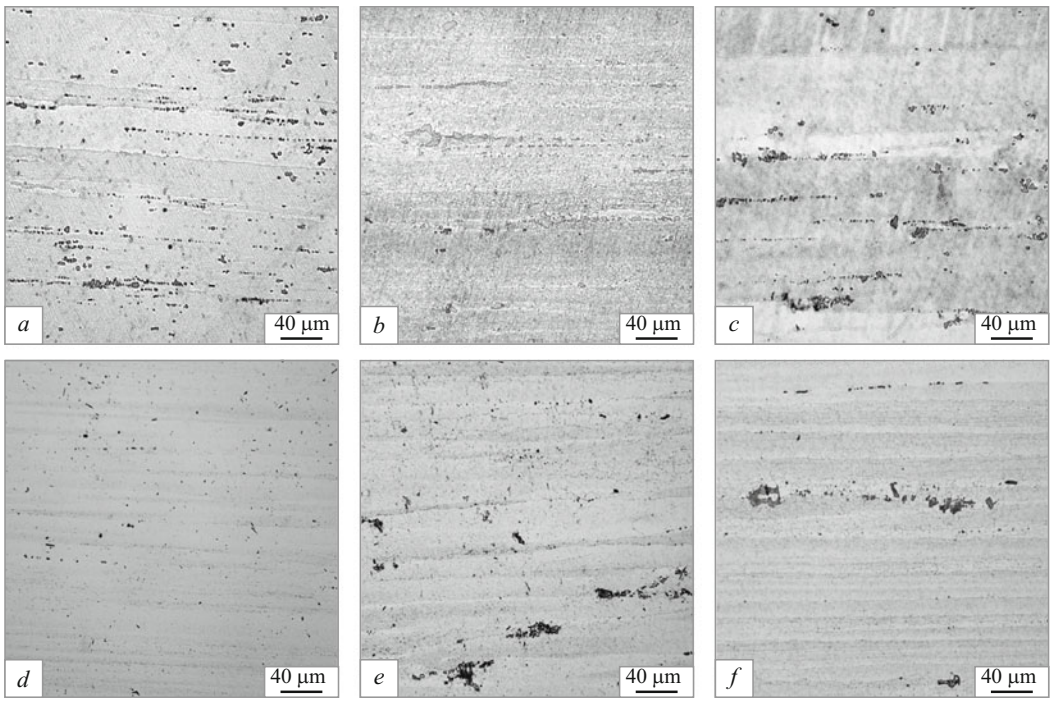

Fig. 3. Microstructure of alloys D16T (*a* – *c*) and 1953T1 (*d* – *f*) as delivered (*a*, *d*) and after 6-min process heating at 250°C followed by operational heating at 150°C for 500 h (*b*, *e*) and 1000 h (*c*, *f*). × 200.

mechanical properties of alloy 1953T1 are more sensitive to process and operational heatings of drill pipes and this results in their substantial softening. The decrease in the values of σ_r and $\sigma_{0.2}$ of alloy 1953T1 after process heating and subsequent operational heating by the regimes studied amounts to 30 – 40% of the values in the state as delivered.

The metallographic and x-ray diffraction studies allowed us to establish that the structure of alloy D16T both as-delivered and after process and operational heatings of different durations contained an aluminum-base α-solid solution and Al_2CuMg (phase *S*) and Al_2Cu (phase θ) intermetallics. This means that the phase composition of the alloy remained unchanged after these heatings, but the experimental data show that their temperature influenced considerably the content and the sizes of the intermetallic inclusions (Fig. 3*a* – *c*).

Table 3 presents the results of a quantitative metallographic analysis of the structure of alloy D16T in the state as-delivered and after process and operational heatings performed

TABLE 3. Content and Sizes of Intermetallic Phases (IP) in the Structure of Alloy D16T As-Delivered and after Production Process and Operational Heatings by Different Regimes

Structural characteristics of IP	State of material											
	Initial state (as delivered)	After process and operational heatings by regimes:										
		200°C						250°C					
		6 min			10 min			6 min			10 min		
		150°C						150°C					
		0 h	500 h	1000 h	0 h	500 h	1000 h	0 h	500 h	1000 h	0 h	500 h	1000 h
Volume fraction, %	3.40	2.53	4.15	5.45	4.16	5.47	6.28	2.89	4.76	5.75	4.30	5.98	6.40
Distribution density, 1/mm^2	12,409	7820	16,366	15,252	14,238	28,125	13,709	8203	16,239	12,110	16,350	24,351	12,648
Mean length, μm	1.80	2.28	1.32	1.56	1.84	1.12	2.32	2.77	1.30	1.51	1.50	1.21	2.77
Mean area, μm^2	2.70	5.61	1.54	1.60	2.50	1.24	4.73	7.35	1.47	1.90	2.80	1.35	5.60
Maximum diameter, μm	2.00	2.19	2.02	2.11	2.60	2.64	2.76	2.35	2.29	2.37	2.77	2.85	3.09

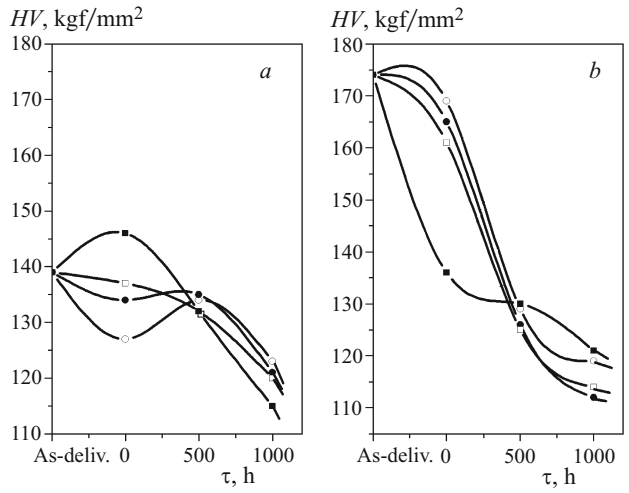

Fig. 4. Effect of the duration of the hold at 150°C on microhardness of alloys D16T (*a*) and 1953T1 (*b*) after preliminary heating (point "0") to 200°C (○, ●) or 250°C (□, ■) with a hold of 6 min (○, □) or 10 min (●, ■).

by various regimes. It can be seen that the 500-h hold at 150°C increases the volume fraction and the distribution density of the intermetallic phases. The mean values of the length and of the area of the inclusions decrease considerably but their maximum diameter is virtually invariable. When the duration of the operational heating is increased from 500 to 1000 h, the quantitative characteristics of the microstructure of the alloy change in the opposite direction except for the volume fraction of the phases; the maximum diameter of the inclusions increases somewhat. Growth in the temperature of the preliminary process heating from 200 to 250°C intensifies somewhat the effect of the operational heating on the determined changes in the structure of the alloy.

Thus, the experimental data obtained show that a hold for up to 500 h at the operational temperature of 150°C increases considerably the total content of secondary intermetallic phases in the structure of D16T due to the growth in the content of fine inclusions. When the hold is prolonged to 1000 h, the intermetallic inclusions coarsen primarily due to dissolution of fine particles.

The regular features determined allow us to assume that in the initial period of the hold at the operational temperature the structure of alloy D16T undergoes precipitation hardening the start of which has been prepared by the preceding process heating [5]. For this reason the structural changes in the alloy at 150°C are intensified somewhat when the duration of the process heating is prolonged.

Growth in the content of fine inclusions in the structure of D16T due to the initial stages of the artificial aging during the hold at 150°C promotes strengthening of the matrix solid solution [6, 7]. A hold of up to 500 h affects little the value of σ_r and virtually does not change the value of $\sigma_{0.2}$ raising somewhat the values of δ and ψ. This fact and the inconsiderable change in the microhardness of the alloy (Fig. 4) reflect preservation of the strength of the matrix. Prolongation of the hold at 150°C to 1000 h is accompanied by coarsening of the secondary intermetallics and causes gradual softening of the alloy [8]. However, this process develops in alloy D16T slowly and does not result in considerable variation of the mechanical properties.

Just like in alloy D16T, operational heating of 1953T1 does not change its phase composition qualitatively but affects substantially the content and size of the secondary inclusions (See Fig. 3*d* – *f*). Both in the state as-delivered and after process and operational heatings the structure of the alloy contains an aluminum-base α-solid solution and $MgZn_2$ and $Al_2M_3Zn_3$ (a *T*-phase) intermetallics.

A quantitative metallographic analysis has shown (Table 4) that in the hold at the operational temperature of 150°C the volume fraction of the intermetallic phases in the

TABLE 4. Content and Sizes of Intermetallic Phases (IP) in the Structure of Alloy 1953T1 As-Delivered and after Process and Operational Heatings by Different Regimes

Structural characteristics of IP	State of material											
	Initial state (as delivered)	After process and operational heatings by regimes:										
		200°C						250°C				
		6 min			10 min			6 min			10 min	
		150°C						150°C				
		0 h	500 h	1000 h	0 h	500 h	1000 h	0 h	500 h	1000 h	0 h	500 h	1000 h
Volume fraction, %	0.50	0.60	1.30	1.46	1.60	1.83	1.85	1.69	2.51	2.68	2.15	2.63	2.78
Distribution density, 1/mm^2	1491	1408	742	679	928	684	655	1485	692	634	800	573	536
Mean length, μm	1.90	2.03	3.88	3.90	2.70	4.47	4.71	2.24	4.45	4.73	3.05	4.83	4.92
Mean area, μm^2	3.10	3.26	5.46	5.89	4.40	6.34	6.80	3.39	5.75	6.33	4.79	6.62	7.10
Maximum diameter, μm	2.10	2.17	4.14	4.31	2.49	4.69	4.83	2.53	4.56	4.87	3.46	4.67	4.73

structure of alloy 1953T1 increases inconsiderably, but their sizes grow markedly and the distribution density decreases. The variation of the structure develops monotonically and intensely from the start of the hold and damps somewhat after 500 h. Note that the determined quantitative changes in the structural parameters during holding at 150°C are much more pronounced in alloy 1953T1 upon growth in the temperature and duration of the preceding process heating.

Analysis of the experimental data reflects intense occurrence of the process of artificial aging in alloy 1953T1 from the start of the hold and in the whole of the range of its duration at 150°C. The aging occurs in the stage of decomposition of the supersaturated α-solid solution with segregation of incoherent secondary phases [6 – 9], which is a factor determining the properties of the alloy, especially in the initial period of the hold. When the hold is extended over 500 h, the alloy overages, and the process damps down progressively. Coarsening of the intermetallics and softening of the matrix solid solution (Fig. 4) cause decrease in the strength characteristics and increase in the ductility characteristics of alloy 1953T1 and are responsible for the lowering of the strain-hardening factor. As a consequence, an operational heating lasting for up to 500 h causes marked softening of alloy 1953T1; prolongation of the hold makes the decrease in the strength less considerable as well as the increase in the ductility of the alloy.

The difference in the kind and intensity of the occurrence of processes in alloys D16T and 1953T1 under the studied heating modes explains their different resistance to fracture in the production and operation of drill pipes. The precipitation hardening and the high structural stability of alloy D16T observed in production process and operational heatings provides inconsiderable changes in their mechanical properties and a high operating capacity. The fast decomposition of the matrix solid solution and the coarsening of the particles of the intermetallic phases in the structure of alloy 1953T1 under production process and operational heatings are responsible for its marked softening. Being stronger than D16T in the state as delivered alloy 1953T1 is inferior to the former with respect to this parameter even after a short-term heating in the production of drill pipes to say nothing about the longer heatings in operation. For example, the values of σ_r and $\sigma_{0.2}$ of alloy 1953T1 in the initial state are 50 and 180 MPa higher than the same characteristics of alloy D16T. However, after the process heating and subsequent hold at the operational temperature of 150°C for 500 h the ultimate rupture strength and the conventional yield strength of alloy 1953T1 are lower than those of alloy D16T by 3 – 40 and 30 – 60 MPa respectively. Alloy 1953T1 is considerably inferior to D16T where the important operational characteristic of the ductility margin $1 - \sigma_{0.2}/\sigma_r$ is concerned (Fig. 5).

Thus, the results obtained show that alloy D16 is more advantageous than alloy 1953 for the production of drill pipes for the oil and gas industry. Alloy D16 does not require strict regulation of the temperature and time parameters in the production of equipment and exhibits higher operational stability. This makes it, firstly, more adaptable to manufacture and, secondly, more suitable for fabrication of drill pipes serving under more rigid operational conditions than alloy 1953.

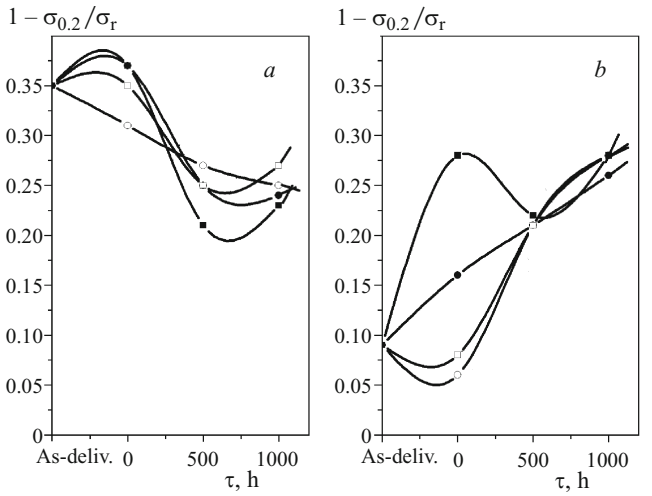

Fig. 5. Effect of the duration of the hold at 150°C on ductility margin ($1 - \sigma_{0.2}/\sigma_r$) of alloys D16T (*a*) and 1953T1 (*b*) after preliminary heating (point "0") to 200°C (○, ●) or 250°C (□, ■) with a hold of 6 min (○, □) or 10 min (●, ■).

CONCLUSIONS

1. We have studied the structural stability and the variation of mechanical properties of drill pipes from aluminum alloys D16T and 1953T1 under production process and operational heatings in the oil and gas industry. Alloy D16T has been shown to be more resistant to the action of heating in the processes of production and operation of pipes than alloy 1953T1.

2. The variation of mechanical properties of alloys D16T and 1953T1 in short-term (10 min) heating to 200 – 250°C and subsequent 1000-h hold at 150°C is caused by the occurrence of diffusion processes in their structure, the kinds of which differ in the alloys studied. In alloy D16T the aging-induced retrogression in short-term heating to 200 – 250°C and the slowly developing precipitation hardening in subsequent long-term heating at 150°C are not accompanied by substantial variation of the strength and ductility. In alloy 1953T1 the fast occurrence of the decomposition of the matrix α-solid solution and the coarsening of the intermetallic particles segregated in the structure cause marked softening.

3. It is necessary to control strictly the temperature and time parameters of the production and the operational parameters of drill pipes from aluminum alloy 1953. Alloy D16 has substantial production of operation advantages over alloy 1953 when these materials are used for making drill pipes for the oil and gas industry.

REFERENCES

1. V. F. Shtamburg, G. M. Fain, S. M. Danelyants, and A. A. Sheina, *Drill Pipes from Aluminum Alloys* [in Russian], Nedra, Moscow (1980), 240 p.
2. S. A. Prokhorov, "Aluminum returns," *Neft' Rossii*, No. 6, 76 – 77 (2011).
3. G. M. Fain and A. X. Neimark, *Design and Operation of Drill Columns for Deep Wells* [in Russian], Nedra, Moscow (1985), 237 p.
4. G. M. Fain, A. A. Golovin, G. A. Voronkov, et al., "Long-term strength of drill pipes from aluminum alloys at elevated service temperatures," *Neft. Prom-st', Ser. Mash. Neft. Oborud.*, No. 8, 10 – 12 (1980).
5. S. Yu. Kondrat'ev and O. V. Shevtsov, "Effect of high-temperature heatings on the structure and properties of aluminum alloys for the production of drill pipes," *Metalloved. Term. Obrab. Met.*, No. 4, 24 – 30 (2013).
6. B. A. Kolachev, V. I. Elagin, and V. A. Livanov, *Metal Science and Heat Treatment of Nonferrous Metals and Alloys* [in Russian], MISiS, Moscow (1999), 413 p.
7. B. A. Kolachev, R. M. Gabidulin, and Yu. V. Piguzov, *Technology of Heat Treatment of Nonferrous Metals and Alloys* [in Russian], Metallurgiya, Moscow (1992), 271 p.
8. I. N. Fridlyander, "Regular features of variation of properties of aluminum alloys during aging," *Metalloved. Term. Obrab. Met.*, No. 9, 8 – 11 (2003).
9. V. S. Muratov and D. P. Yudaev, "Effect of additional aging in production heating on mechanical properties and microstructure of sheet semiproducts from alloy 1151," *Zagotov. Proizvod. Mashinostr.*, No. 11, 41 – 43 (2009).