

THE PARAMETERS OF MARTENSITIC TRANSFORMATIONS IN Ti₃₂Hf₁₈Ni_{50-x}Cu_x ALLOYS

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The multi-component alloys on TiNi basis, where the relation «A(Ti):B(Ni)» 50:50, A:Ti₃₂Hf₁₈Ni_{50-x}Cu_x (where x=0; 5; 15; 25; 35; 45at.%) are elaborated. All alloys are obtained and investigated by different methods. One can reveal, that region of homogeneous B2-solid solution stays constant only at cuprum concentrations, not exceeding 28 at%. The alloys at big concentrations become supersaturated and their eutectic transformation with formation of phase mixB2-TiNi and B11-TiCu take place.

Introduction.

The study of martensitic transformations, to which the big group of structural phase transfers of diffusion-free cooperative type, presents the certain interest. As martensitic transformations in many cases are accompanied by unusual effects, then they open the wide perspectives for practical applications. But from the other hand our task was complicated by the fact, that it is necessary to take into consideration the tendency of alloying element to amorphous transformation at ultraspeed tempering, saving at this the effect of shape memory, hyperelasticity and other functional properties after the crystallization.

The alloy obtaining on the basis of titanium nickelide from liquid state has the set of advantages: the ability of fixing of high-temperature state without additional thermal treatment, prevention of decay of solid solution, ability to regulate of dimension of solid grain and width of obtained tapes in definite limits. The appearance of new structural and phase states is possible at such method of material obtaining.

The solidification of melted material with extreme-high cooling rates (more than 10⁵⁰C/c) allows to obtain practically whole gamma of amorphous and microcrystal materials with enough high productivity. The heat bend from solidificating melt can be carried out by radiation, convection and heat capacity. The concrete mechanism of heat transfer both single (at cooling melt drops in vacuum because of radiation) and combined ones is realized in the dependence on obtaining method.

The choice of material and experiment technique.

The choice of chemical composition for multi-component alloys with the effect of shape memory on TiNi basis, which further will be obtained by ultraspeed tempering from the melt, is based on two principles. Firstly, chemical composition should be related to amorphous transformation at ultraspeed tempering from the melt, secondly the alloy should have the martensitic transformations in massive state (till ultraspeed tempering from the melt).

In our paper the quick-tempered tapes (of foil), obtained from "master" alloys on the technique, described in [1] are investigated, the ultraspeed tempering from melt (flow turning), Planar Flow Casting.

The multi-component alloys on TiNi basis, where the relation «A(Ti):B(Ni)» 50:50 are chosen for the investigation. The A: Ti₃₂Hf₁₈Ni_{50-x}Cu_x type (where x= 0; 5; 15; 25; 35; 45at.%), where only Cu is varied in the composition is elaborated.

The differential scanning calorimeter (DSC) Mettler DSC 30 and Mettler DSC 822e was used for the investigation of thermodynamic properties. These devices alloy to carry out the investigations in temperature range from -150°C till 700°C with heating rate 10-10K/c and cooling rate 10-20 K/c. The sample masses are from 5 till 10 mg, box materials, in which the samples are situated, are platinum and aluminum ones. The main complex of experimental investigation methods is also applied: X-ray diffraction analysis, resistivities, spectroscopy, transmission and raster electron microscopy.

The nominal chemical composition of investigated quick-tempered tapes of A type (Ti, Hf)(Ni, Cu), melt temperature at T_q output and width of obtained foils d are given in the table 1.

Table 1

The nominal chemical composition of quick-tempered foils

Alloy	Nominal chemical composition (at. %)				$T_q(^{\circ}C)$	d (mm)
	Ti	Hf	Ni	Cu		
A0	32	18	50	-	1380	30
A1	32	18	45	5	1296	30
A2	32	18	35	15	1285	30
A3	32	18	25	25	1300	30
A4	32	18	15	35	1200	30
A5	32	18	5	45	1100	30

Table 2

The main characteristics of martensitic transformation in elaborated system alloys (Ti,Hf)(Ni,Cu) in massive state after obtaining.

Alloy	Phase transformations	Parameters of martensitic transformations						
		M_s	M_p	M_f	A_s	A_p	A_f	
C°								
A1	B2↔(B19+B19')	161	122	4	50	178	226	
A2	B2→B19		-8	-	120	150	181	
	B19→B19'	147	-46	12	-35	2	74	
A3	B2→B19			-				
	B19↔B19'	121	57	-28	1	78	175	
		-140	-155	<	-	-69	8	
A4	There are no martensitic transformations in temperature interval from -196 till +250°C.							
A5								

All these alloys firstly are obtained by arc melting (sixfold remelt), by the technique, described in [2]. The martensitic transformations are revealed not in all alloys. The main characteristics of martensitic transformation in

elaborated system alloys (Ti,Hf)(Ni,Cu) in massive state after obtaining are given in table 2, main characteristics after homogenization at temperature 950°C during 1 hour, including ΔT (MP hysteresis) are given in table 3.

The results and their discussion.

One can reveal in studied cuprum alloys of A type: $Ti_{32}Hf_{18}Ni_{25}Cu_{25} \rightarrow Ti_{32}Hf_{18}Ni_{50-x}Cu_x$ (where $x=5 \div 45$ at.%), that region of homogeneous B2-solid solution saves only at cuprum concentration, not exceeding 28at.%. At big concentrations the cuprum alloys become supersaturated and the eutectic transformation with formation of phase mix B2-TiNi and B11-TiCu take place. The alloy $Ti_{32}Hf_{18}Ni_{25}Cu_{25}$, chosen by us, is in the region of solid solution (table 2 and 3).

As it is seen from the tables, both martensitic phases in alloys with small cuprum additions coexist in the one temperature interval, where M_s is 161°C, A_f is 226°C; these two transitions after homogenized annealing at temperature 900°C during 1 hour, and parameters of martensitic transformation are some increased correspondingly till 190 and 236°C.

The phase transformations in alloys with 15 at.% of cuprum carry out in two stages: on the first stage $B2 \leftrightarrow B19'$, and on the second one $B19 \leftrightarrow B19'$. The parameters of martensitic transformations are essentially changed: M_s

($B2 \leftrightarrow B19$) decreases till 147°C, and A_f till 180°C. The parameters of martensitic transformation $B2 \leftrightarrow B19$ are some increased $M_s=178^\circ C$, $A_f=200^\circ C$ after analogous thermal treatment (homogenization at 900°C during 1 hour). The parameters of second transition are also changed $B19 \leftrightarrow B19'$: before homogenization $M_s=-12^\circ C$, $A_f=74^\circ C$, after homogenization $M_s=72^\circ C$, $A_f=130^\circ C$.

The two martensitic transitions $B2 \leftrightarrow B19$ and $B19 \leftrightarrow B19'$ are also observed in $Ti_{32}Hf_{18}Ni_{25}Cu_{25}$ alloy, but parameters of martensitic transformations decrease till 121°C (M_s) and till 175°C (M_f , $B2 \leftrightarrow B19$), after homogenization M_s corresponds to 100°C, and $M_f=250^\circ$. The second transformation $B19 \leftrightarrow B19'$ goes deeper (for example, M_f ($B19 \leftrightarrow B19'$) corresponds to temperature below -196°C), after thermal treatment one can stabilize and increase M_f ($B19 \leftrightarrow B19'$) till temperature -185°C. The alloy with 25 at.% Cu on its main parameters is the more interesting for further investigation.

The martensitic transformation in temperature interval -196÷+250°C one can't fix in $Ti_{32}Hf_{18}Ni_{50-x}Cu_x$ ($x=35 \div 45$ at.%) alloys. The martensitic transformation isn't also revealed after homogenized annealing at 900°C during several hours.

Table 3

The main characteristics of martensitic transformation in elaborated alloys of (Ti,Hf)(Ni,Cu) systems in massive state after homogenization at temperature 950°C during 1 hour (ΔT is MP hysteresis).

Alloy	Characteristics of martensitic transformation	M_s ,	M_p ,	M_f ,	A_s ,	A_p ,	A_f ,	ΔT ,
		°C	°C	°C	°C	°C	°C	
A0	$B2 \leftrightarrow B19'$	230	212	150	225	258	287	55
A1	$B2 \leftrightarrow (B19+B19')$	190	130	0	61	192	236	62
A2	$B2 \leftrightarrow B19$ $B19 \leftrightarrow B19'$	178	130	72	145	170	200	20
		72	30	-48	25	70	130	55
A3	$B2 \leftrightarrow B19$ $B19 \leftrightarrow B19'$	121	57	-28	1	78	175	20
		-143	-155	<-175	-99	-58	1	>100

The typical curves of differential scanning calorimetry at temperature -150÷600°C and cyclic cooling process ↔ heating for $Ti_{32}Hf_{18}Ni_{45}Cu_5$ and $Ti_{32}Hf_{18}Ni_{35}Cu_{15}$ alloys more detail reveal the character and consistency of martensitic transformations. The diagram of martensitic transformations for four-component alloy TiHfNiCu is similar on metastable diagram of triple alloys TiNiCu (fig.1).

But temperatures of martensitic transformations are shifted to more high temperatures, that makes them more interesting ones.

The character of temperature dependencies of beginning of first and second martensitic transitions from Cu composition in $Ti_{32}Hf_{18}Ni_{50-x}Cu_x$ alloys in the comparison with triple alloys TiNiCu is shown on fig.2. The given phase diagram is obtained by us for massive samples of alloys of TiHfNiCu system in initial state and after homogenized annealing 900°C during 1 hour. The metastable phase diagram of triple alloy TiNiCu, published in previous papers [3,4] for the comparison is given on the fig.1. From the fig.1 it is seen, that hafnium addition instead of titanium 18at.%

alloys the elaboration of four-component alloys with the effect of shape memory in the wide temperature range.

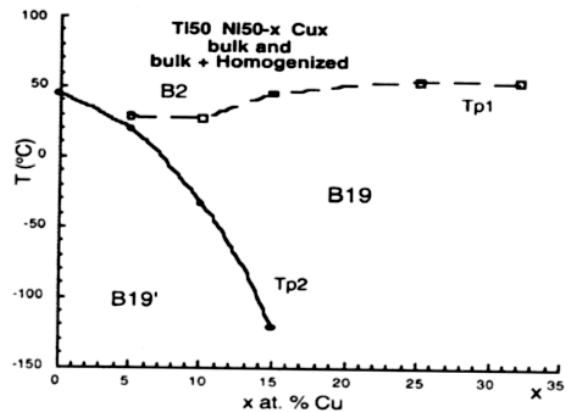


Fig.1. The diagram of metastable states of $Ti_{50}Ni_{50-x}Cu_x$ ($p < X < 35$ at.%) alloy initial state + homogenized annealing 900°C, 1 hour.

The generalized data on glass transformation heating T_g , crystallization T_x , and also $\Delta T = T_x - T_g$ and crystallization heating ΔH for investigated multi-component alloys of A type of obtained ultraspeed tempering from the melt. The

foils have different structural states after ultraspeed tempering (in initial state), for example A0 and A1 alloys are in shifted (amorphously-crystal), and A2 and A5 alloys are totally in amorphous state are given in the table 4.

Table 4

T_g , T_x , ΔT temperatures and also temperature of crystallization heating of ΔH amorphous phase of quick-tempered foils of (Ti,Hf)(Ni,Cu) and (Ti,Hf,Zr)[(Ni,Co)(Cu,Ag)] systems, the data are obtained at heating rate 20 K/m.

Alloy calling and nominal chemical composition, at. %	Relations A:B	T_g , °C	T_x , °C	$\Delta T = (T_x - T_g)$, °C	ΔH , J/g
A0-Ti ₃₂ Hf ₁₈ Ni ₅₀	50:50	480	520	40	7
A1-Ti ₃₂ Hf ₁₈ Ni ₄₅ Cu ₅	50:50	483	511	28	16
A2-Ti ₃₂ Hf ₁₈ Ni ₃₅ Cu ₁₅	50:50	456	500	44	17
A3-Ti ₃₂ Hf ₁₈ Ni ₂₅ Cu ₂₅	50:50	490	515	25	27
A4-Ti ₃₂ Hf ₁₈ Ni ₁₅ Cu ₃₅	50:50	459	488	30	33
A5-Ti ₃₂ Hf ₁₈ Ni ₅ Cu ₄₅	50:50	435	474	40	29

The vitrification temperature T_g is the one of the more important characteristics of amorphous metals and alloys. It is fixed in the form of small endothermic peak preceding to crystallization beginning on the curves of differential thermal analysis. The temperature T_x , by definition, is temperature of crystallization beginning of amorphous phase at heating. It is accepted, that small volume (10^{-4}) of crystal phase appears in the crystallization beginning. T_x and T_g . That's why it is necessary to take into consideration at which velocity these values have been obtained at comparison of values of these temperatures. The interval $\Delta T = T_x - T_g$ is called by temperature interval of supercooled liquid [4-7].

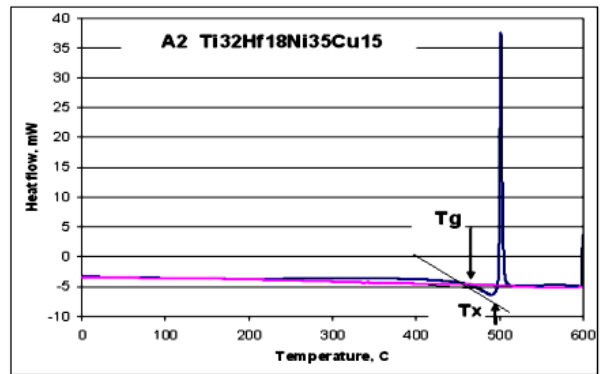


Fig.3. The calorimetric curve of quick-tempered foils of A2(Ti₃₂Hf₁₈Ni₃₅Cu₁₅) alloy in the initial state (first and the second cycle of heating), heating rate is 20 K/min.

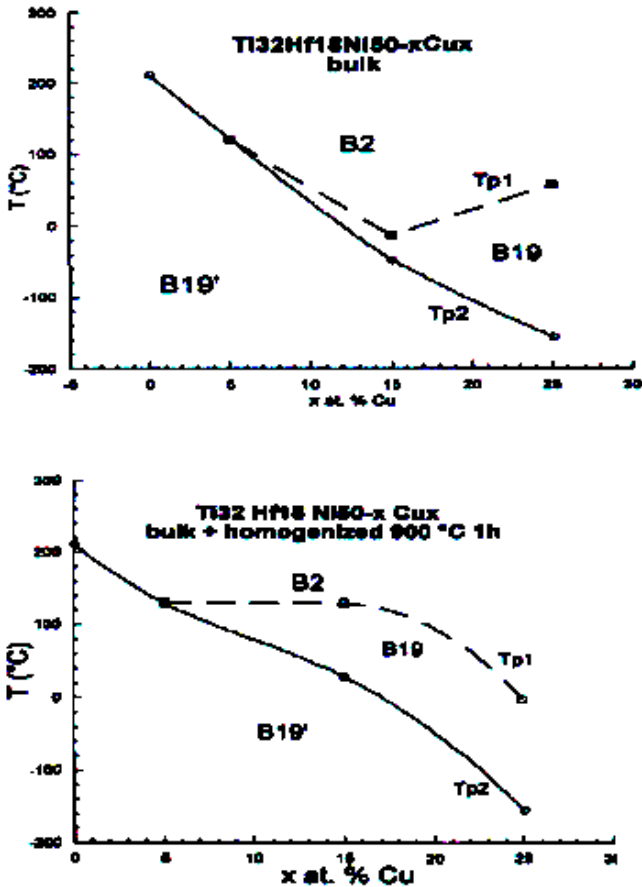


Fig.2. The diagram of metastable states of Ti₃₂Hf₁₈Ni_{50-x}Cu_x (0 < x < 45 at. %) alloy a) in initial state; b) after homogenized annealing 900°C, 1 hour.

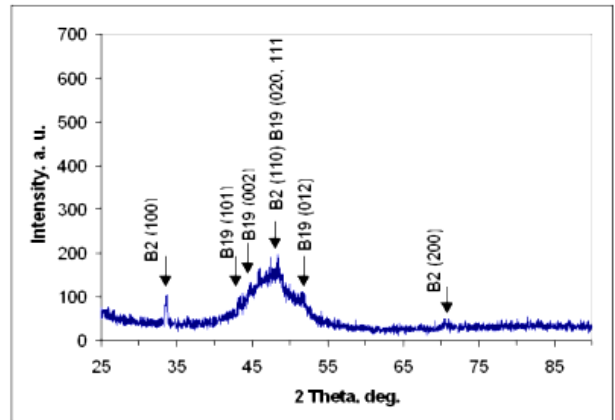


Fig.4. The diffractogram of quick-tempered foils of A1 (Ti₃₂Hf₁₈Ni₄₅Cu₅) alloy in the initial state at room temperature ("free" foil surface).

The foils with amorphous structure at heating demonstrate the set of peculiarities. These changes can be fixed by calorimeter investigations. The typical curve for A2 alloy in initial state is given on the fig.3. The weak variation of thermal current (exothermic reaction) is fixed at heating in the interval from $T = 200^\circ\text{C}$ till T_g (glass transition temperature), that is connected with carrying out of structural relaxation in amorphous phase. The transition through vitrification temperature is reversible one, i.e. amorphous metal, heated higher than T_g , but below than crystallization temperature it can be cooled with the formation of amorphous

structure. The set of thermal treatments in the following temperature intervals is carried out for the investigation of structural changes at heating of quick-tempered foils of multi-component alloys on NiTi basis with amorphous structure: below glass vitrification temperature T_g ; in interval $T_x - T_g$; higher than crystallization temperature T_x , that allows to consider all possible variants of microstructure formation in investigated alloys of A type.

The results of X-ray diffraction phase analysis of $Al(Ti_{32}Hf_{18}Ni_{45}Cu_5)$ alloy in initial state "after obtaining" at room temperature is presented on the fig.4. The diffractogram decryption shows, that there are amorphous and two crystal phases: B2 austenite and B19 martensite, that's why the given alloy is chosen for further investigations as model one. The structure evolution at heating is more detail investigated on the example of the given alloy.

The data, obtained with the help of transmission electron microscopy for alloy A1 ($Ti_{32}Hf_{18}Ni_{45}Cu_5$) in the initial "after obtaining" state at room temperature (bright-field image and electron diffraction) are presented on the fig.5. They prove the results of X-ray diffraction phase analysis and evidence about the fact, that amorphous and two crystal phases: B2 austenite and B19 martensite, indeed coexist in A1 in initial state at room temperature.

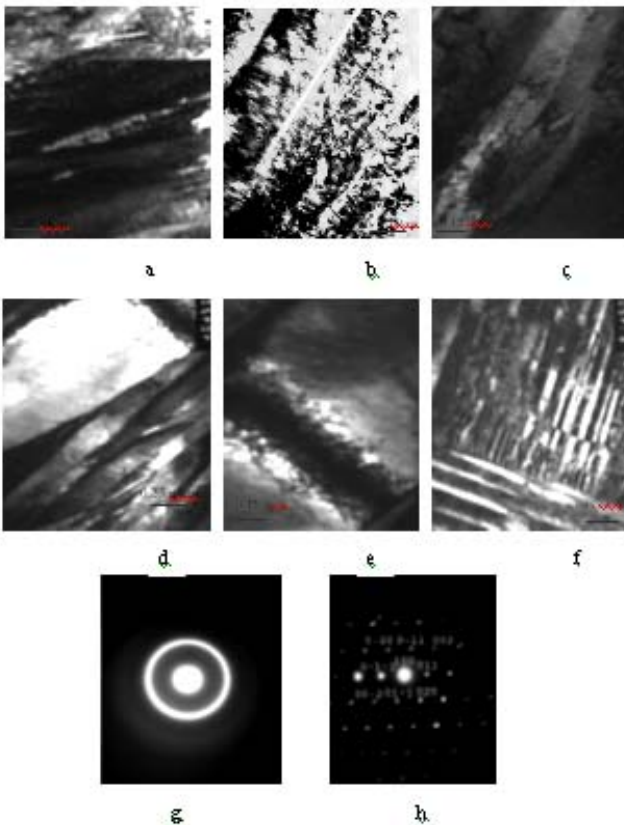


Fig.5. The bright-field a-f and electron diffraction g (amorphous and crystal phases), h (B19 martensite) the films, obtained with the help of PEM for $Al(Ti_{32}Hf_{18}Ni_{45}Cu_5)$ alloy in the initial state; a, b and c are B19 martensite plates (by the width $\sim 0,3-0,4$ μm); d is B19 martensite plates (bt the width $\sim 0,3-0,4$ μm , thin doubles; e is dislocation accumulation in martensitic phase; f is self-accommodated groups of martensitic plates.

The martensite has the plate morphology, moreover plate dimensions vary in wide interval from 0,05 till 0,73 μm . The following set of thermal treatments is carried out for this

alloy: 457°C, 2 min.; 467°C, 2 min.; 477°C, 2 min.; 487°C, 2 min.; 502°C, 2 min. at temperatures below T_g and in temperature interval $[T_g \div T_x]$. After it the samples are investigated with the help of calorimeter and the information relatively thermal stability of amorphous phase after carried out thermal treatments and characteristics of martensitic transformation in formed grains of crystal phase are obtained. The investigations are carried out on the following scheme: 20°C \rightarrow 200°C \rightarrow -150°C \rightarrow 600°C \rightarrow -150°C \rightarrow 600°C, cooling and heating rates $T' = 20$ K/min. The calorimeter curves are given on the fig.6-13.

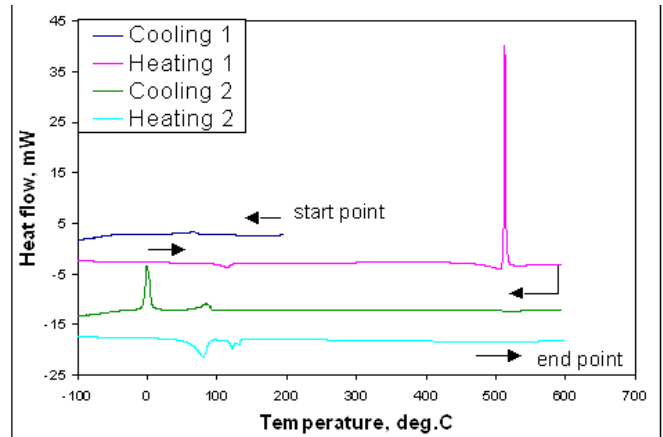


Fig.6. The calorimetric curves of rapid-tempered foils of $Al(Ti_{32}Hf_{18}Ni_{45}Cu_5)$ alloy in the initial state (upper curves are peaks of direct and reversal martensitic transformations $B2 \leftrightarrow B19$; lower ones are additional peaks of martensitic transformation $B2 \leftrightarrow B19'$ after heating till 600°C.

From the given data it is followed, that thermal treatment of the given alloy in temperature interval till T_g doesn't influence on crystallization processes of amorphous phases and on carrying out of the martensitic transformation both before and after crystallization of initial amorphous phase. The process of partial crystallization of amorphous part of structure (value ΔH after preliminary thermal treatment decreases) at heating of crystallization temperature in $[T_g \div T_x]$ interval takes place.

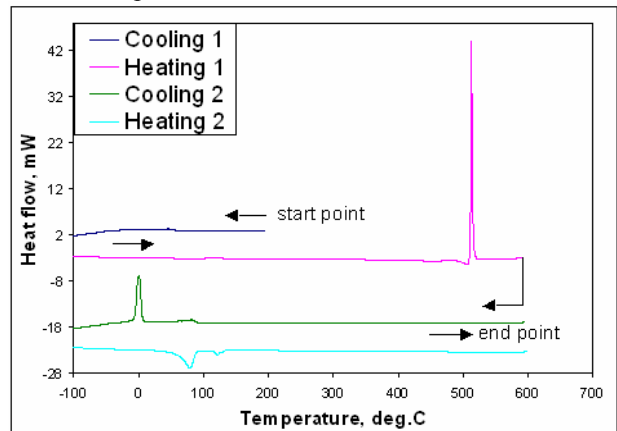


Fig.7. The calorimetric curves of rapid-tempered foils of $Al(Ti_{32}Hf_{18}Ni_{45}Cu_5)$ alloy after preliminary thermal treatment at 457°C, 2 min. (upper curves are $B2 \leftrightarrow B19$ transformations; lower ones are $B2 \leftrightarrow B19 \leftrightarrow B19'$ after heating till 600°C).

The investigation results (X-ray diffraction phase analysis and transmission electron microscopy) of these samples are presented on the fig.14-18.

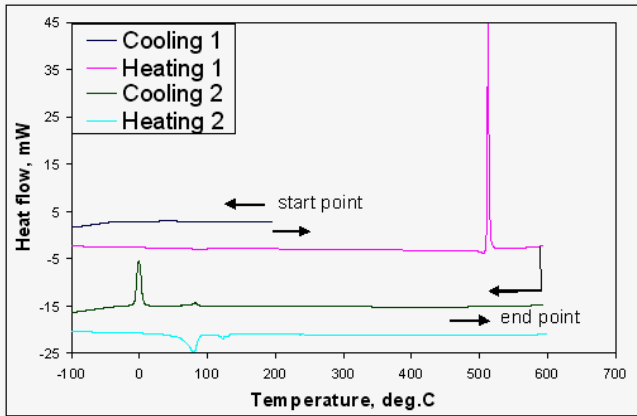


Fig.8. The calorimetric curves of quick-tempered foils of A1 ($Ti_{32}Hf_{18}Ni_{45}Cu_5$) alloy after preliminary treatment at 467°C, 2 min. (upper curves are transformations $B2 \leftrightarrow B19$; lower ones are $B2 \leftrightarrow B19 \leftrightarrow B19'$ after heating till 600°C).

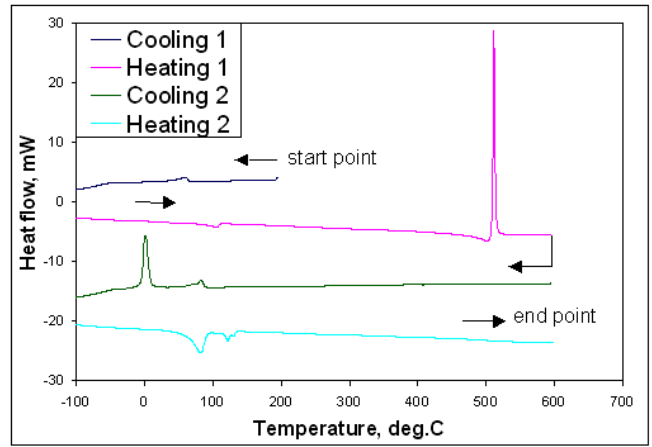


Fig.11. The calorimetric curves of quick-tempered foils of A1 ($Ti_{32}Hf_{18}Ni_{45}Cu_5$) alloy after preliminary thermal treatment at 497°C, 2 min. (upper curves are transformations $B2 \leftrightarrow B19$; lower ones are $B2 \leftrightarrow B19 \leftrightarrow B19'$ after heating till 600°C).

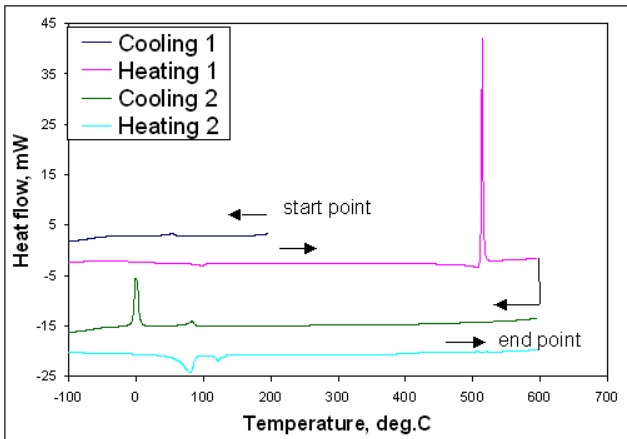


Fig.9. The calorimetric curves of quick-tempered foils of A1 ($Ti_{32}Hf_{18}Ni_{45}Cu_5$) alloy after preliminary thermal treatment at 477°C, 2 min. (upper curves are transformations $B2 \leftrightarrow B19$; lower ones are $B2 \leftrightarrow B19 \leftrightarrow B19'$ after heating till 600°C).

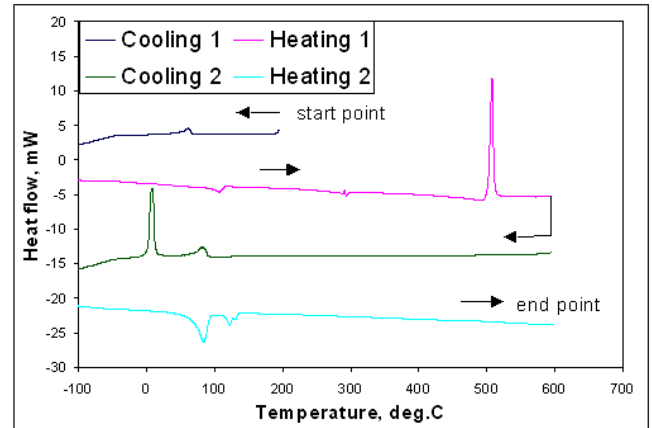


Fig.12. The calorimetric curves of quick-tempered foils of A1 ($Ti_{32}Hf_{18}Ni_{45}Cu_5$) alloy after preliminary thermal treatment at 502°C, 2 min. (upper curves are transformations $B2 \leftrightarrow B19$; lower ones are $B2 \leftrightarrow B19 \leftrightarrow B19'$ after heating till 600°C).

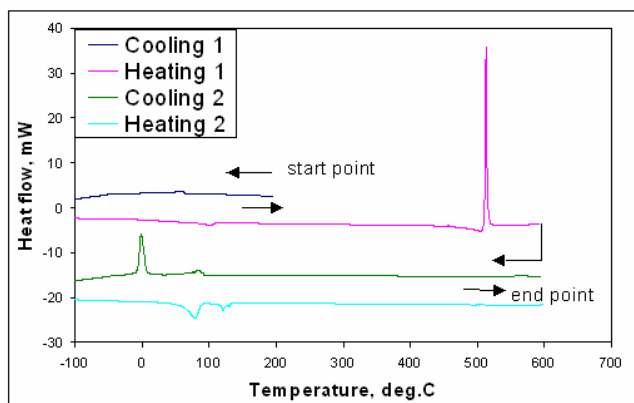


Fig.10. The calorimetric curves of quick-tempered foils of A1 ($Ti_{32}Hf_{18}Ni_{45}Cu_5$) alloy after preliminary thermal treatment at 487°C, 2 min. (upper curves are transformations $B2 \leftrightarrow B19$; lower ones are $B2 \leftrightarrow B19 \leftrightarrow B19'$ after heating till 600°C).

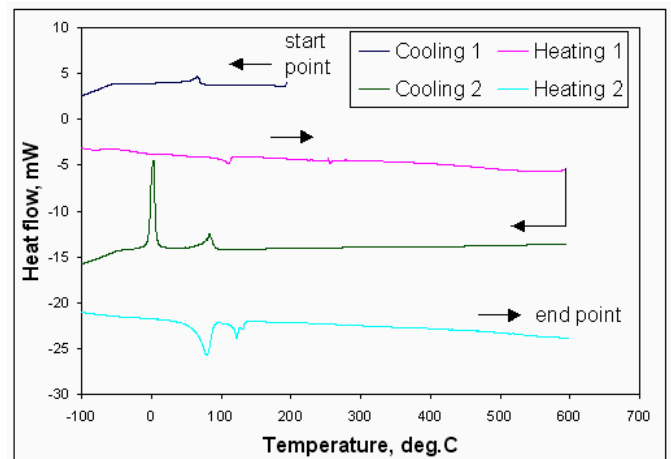


Fig.13. The calorimetric curves of quick-tempered foils of A1 ($Ti_{32}Hf_{18}Ni_{45}Cu_5$) alloy after preliminary thermal treatment at 507°C, 2 min. (upper curves are transformations $B2 \leftrightarrow B19$; lower ones are $B2 \leftrightarrow B19 \leftrightarrow B19'$ after heating till 600°C).

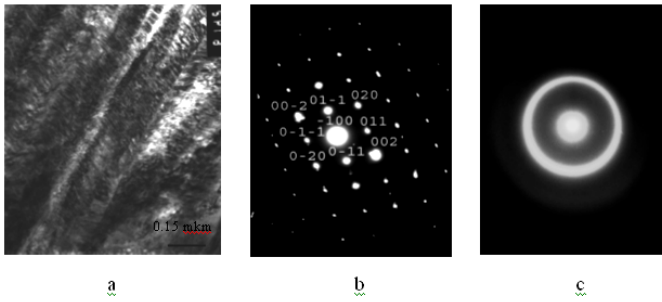


Fig.14. The martensite microstructure B19 (bright-field image (a)) and corresponding electron diffraction (b), and also diffraction from residual amorphous phase (c) of A1 alloy foils after thermal treatment at 497°C, 2 min.

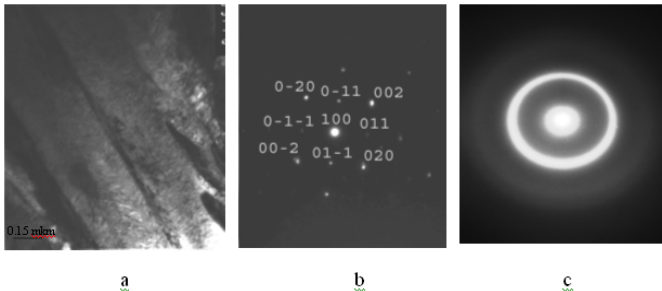


Fig.15. The martensite microstructure B19 (bright-field image (a)) and corresponding electron diffraction (b), and also diffraction from residual amorphous phase (c) of A1 alloy foils after thermal treatment at 502°C, 2 min.

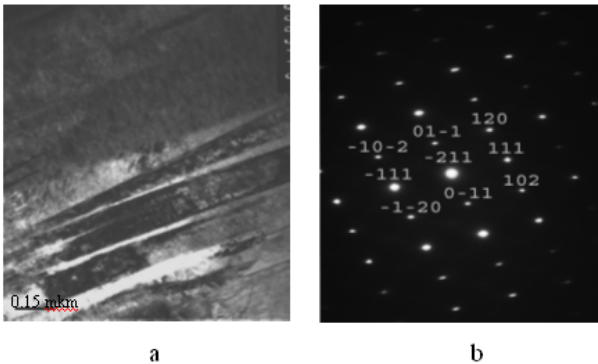


Fig.16. The martensite microstructure B19 (bright-field image (a)) and corresponding electron diffraction (b) of A1 alloy foils after thermal treatment at 507°C, 2 min.

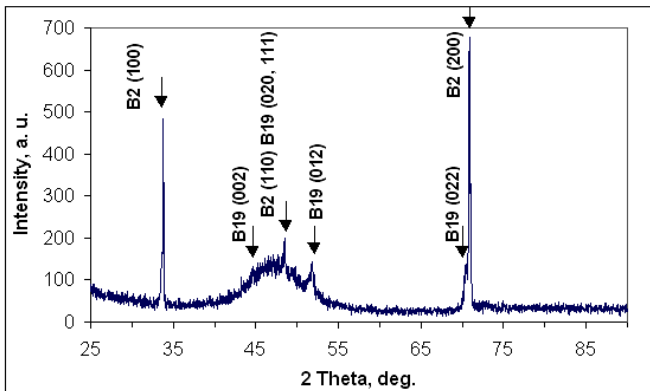


Fig.17. The diffractogram of quick-tempered foils of A1($Ti_{32}Hf_{18}Ni_{45}Cu_5$) alloy after preliminary thermal treatment at 502°C, 2 min (“free” foil surface).

The obtained results of complex investigation of structural changes at heating of multi-component alloys on TiNi basis of A type: (Ti, Ni)(Ni,Cu) allow to construct the evolution of the structure for following positions:

1. In the case, when foils have initially amorphous structure the alloy heating from room temperature till glass transition temperature T_g doesn't lead to crystal formation in amorphous matrix and other changes of phase composition isn't observed. But by the other hand it can influence on processes of structural relaxation of amorphous alloys [8,13] during heating and at delay of the given temperatures below T_g .

Some authors instead the term “ T_g vitrification temperature” use the term “ T_f dummy temperature”, but physical meaning of these terms is similar.

At the same time the alloy heating with amorphous structure in $[T_g-T_x]$ interval leads to creation and growth of crystal phase in amorphous matrix. The type of crystal phase B2 is ordered by body-centered cubic structure of CsCl type. The dimension of once again formed crystals depends on heating temperature and time of isothermic delay at these temperatures.

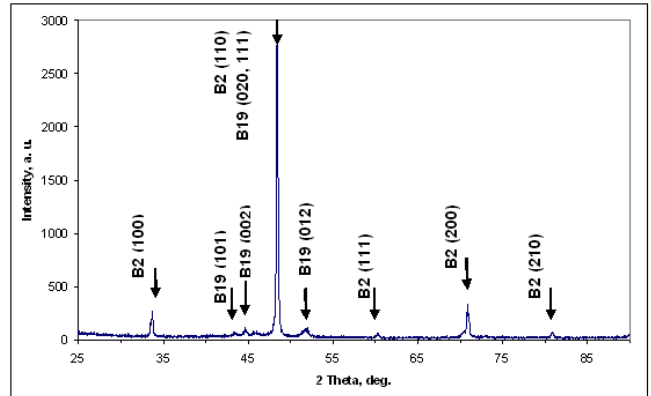


Fig.18. The diffractogram of quick-tempered foils of A1($Ti_{32}Hf_{18}Ni_{45}Cu_5$) alloy after preliminary thermal treatment at 507°C, 2 min (“free” foil surface).

In alloys, where the more than one crystallization peak is fixed at calorimetric investigations, then this result is explained by the process of phase off-orientation in amorphous alloys.

2. In the case, when foils have mix structure (amorphous + crystal phases), the alloy heating leads to the following results: the behavior of amorphous structure carries out analogically to actions, described in point 1. If we take into consideration the fact, that there is crystal phase besides amorphous one (let's call it crystal phase №1), then structure evolution at heating has the following form (fig.19):

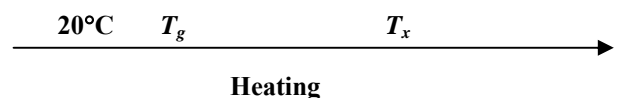
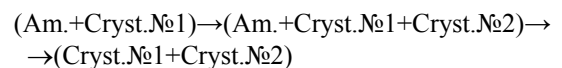


Fig.19. The scheme of structure evolution at heating. The alloy with shifted initial structure: amorphous and crystal phases.

For the final revealing of the presence of martensitic transformation into quick-tempered foils of investigated alloys of A type we carry out the one more set of thermal treatments. The obtained results are given in the table 5 (the more interesting results are taken: one alloy (A2-Ti₃₂Hf₁₈Ni₃₅Cu₁₅) has the initially amorphous structure, and other alloy (A1-Ti₃₂Hf₁₈Ni₄₅Cu₅) has the shifted structure: amorphous and crystal phases). In that alloys (A4 and A5), where there is no martensitic transformation in massive state one can't fix the martensitic transformations. These alloys transform into crystal state, but they don't have martensitic transformations at thermal treatment at 900°C during 1 hour.

Conclusion.

Thus, analyzing the obtained data, we can make several conclusions:

1. Not all elaborated multi-component alloys on TiNi basis, where relation «A(Ti):B(Ni)» 50:50 of A:Ti₃₂Hf₁₈Ni_{50-x}Cu_x type (where x=0; 5; 15; 25; 35; 45 at.%) show the martensitic transformations in initially massive state (after arc smelting). One can't fix the martensitic transformation in A4 and A5 alloys even after homogenized annealing at temperature 900°C during 1 hour.

2. All investigated multi-component alloys if A type are obtained of ultra-speed tempering from the melt. The some foils in the initial state ("after obtaining") have shifted structure (amorphous and crystal phases), these are A0 and

A1 alloys. But other alloys (A2-A5) have totally amorphous structure in the initial state.

3. The type of martensitic phase at room temperature in the initial state ("after obtaining") in A1 alloy (Ti₃₂Hf₁₈Ni₄₅Cu₅), and also after thermal treatments in temperature interval 457°C-507°C is following: B19 (orthorhombic one).

4. The same martensitic phase (B19) has the plate morphology at room temperature in the initial state ("after obtaining") and doesn't endure any change after thermal treatments in temperature interval 457°C-507°C.

5. The quick-tempered foils of A1 alloy (Ti₃₂Hf₁₈Ni₄₅Cu₅) endure the martensitic transformation B2↔B19 in the initial state ("after obtaining") and after thermal treatment in T_g - T_x interval. The alloy has the martensitic transformation B2↔B19↔B19' after heating till 600°C. This change can be explained by the increase of austenite grain (dimension effect) (crystal phase №1 – B2 phase) and by formation of crystal phase of №2 – B2 structure or by alloy heating in the region of stable β-phase B2.

6. The thermal treatment in the temperature interval T_g-T_x doesn't practically change the temperature of martensitic transformation in A1 (Ti₃₂Hf₁₈Ni₄₅Cu₅) alloy.

7. The martensitic transformation isn't observed in alloys A4-Ti₃₂Hf₁₈Ni₁₅Cu₃₅ and A5-Ti₃₂Hf₁₈Ni₅Cu₄₅, obtained by ultra-speed tempering from the melt and additionally thermally elaborated at temperature 900°C during 1 hour.

Table 5

The main characteristics of martensitic transformation in alloys on NiTi basis of A type: (Ti,Hf)(Ni,Cu) in the form of quick-tempered foils after different modes of thermal treatment.

Alloy	Characteristics of martensitic transformation						
	Transformation type	M _s , °C	M _p , °C	M _f , °C	A _s , °C	A _p , °C	A _f , °C
Thermal treatment at 570°C 1 min							
A1	B2↔B19↔B19'	83	64	20	98	111	132
		1	-9	-39	-8	71	92
A2	-	It is not revealed in temperature interval -150÷600°C					
Thermal treatment at 700°C 1 hour							
A1	B2↔B19↔B19'	87	81	53	113	127	133
		53	33	1	57	98	113
A2	B2↔B19	-51	-56	-59	-6	5	11
Thermal treatment at 800°C 1 hour							
A1	B2↔B19↔B19'	71	64			122	124
			5	-1	60	65	
A2	B2↔B19	-44	-47; -77	-79	-24	-22; 4	7
Thermal treatment at 900°C 1 hour							
A1	B2↔B19↔B19'	112	87			140	146
			64	48	98	119	
A2	B2↔B19	-38	-43	-63	-12	10	13

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$Ti_{32}Hf_{18}Ni_{50-x}Cu_x$ ƏRİNTİLƏRİNDƏ MARTENSİT ÇEVİRİLMƏLƏRİNİN PARAMETRLƏRİ

$TiNi$ əsaslı çoxkomponentli ərintilər «A(Ti):V(Ni)» 50:50 nisbətində $Ti_{32}Hf_{18}Ni_{50-x}Cu_x$ ($x=0; 5; 15; 25; 35; 45$ at.%) işlənmişdir. Bütün ərintilər müxtəlif üsullarla hazırlanmış və tədqiq edilmişdir. Məlum olmuşdur ki, öyrənilmiş ərintilərdə bircinsli B2 bərk məhlul yalnız tərkibdə misin miqdarı 28 at.% -nə qədər olduqda saxlanılır. Ərintidə misin konsentrasiyasını artırırdıqda ifrat doyma halı qeyd olunur və nəticədə evtektik çevrilmə baş verir: V2- $TiNi$ və V11- $TiCu$.

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ПАРАМЕТРЫ МАРТЕНСИТНЫХ ПРЕВРАЩЕНИЙ В СПЛАВАХ $Ti_{32}Hf_{18}Ni_{50-x}Cu_x$

Разработаны многокомпонентные сплавы на основе $TiNi$, где соотношение «A(Ti):B(Ni)» 50:50, серия A: $Ti_{32}Hf_{18}Ni_{50-x}Cu_x$ (где $x=0; 5; 15; 25; 35; 45$ ат. %). Все сплавы получены и исследованы разными методами. Выяснили, что область однородного B2-твердого раствора сохраняется лишь при концентрациях меди, не превышающих 28 ат.%. При больших концентрациях меди сплавы оказываются пересыщенными и испытывают эвтектическое превращение с образованием смеси фаз B2- $TiNi$ и B11- $TiCu$.

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