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Hot Plasticity of Alpha Beta Alloys

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1. Introduction

Two phase titanium alloys most often are hot deformed, mainly by open die or close-die forging. Desired mechanical properties can be achieved in these alloys by development of proper microstructure in plastic working and heat treatment processes. Irreversible microstructural changes caused by deformation at the temperature in $\alpha+\beta\leftrightarrow\beta$ phase transformation range quite often cannot be eliminated or reduced by heat treatment and therefore required properties of products cannot be achieved (Bylica & Sieniawski, 1985; Lütjering, 1998; Zwicker, 1974). Some of the properties of titanium alloys, such as: high chemical affinity to oxygen, low thermal conductivity, high heat capacity and significant dependence of plastic flow resistance on strain rate, make it very difficult to obtain finished products having desired microstructure and properties by hot working. Differences in temperature across the material volume, which result from various deformation conditions (local strain and strain rate) lead to formation of zones having various phase composition (equilibrium α and β phases, martensitic phases α' (α'')), morphology (equiaxial, lamellar, bimodal) and dispersion (fine- or coarse-grained) and therefore various mechanical properties (Kubiak & Sieniawski, 1998).

Obtaining desired microstructure of Ti-6Al-4V titanium alloy using plastic deformation in the $\alpha+\beta\leftrightarrow\beta$ phase transformation range is related to proper conditions selection taking into account plastic deformation, phase transformation, dynamic recovery and recrystallization effects (Ding et al., 2002; Kubiak, 2004; Kubiak & Sieniawski, 1998). Grain refinement can be achieved by including preliminary heat treatment in thermomechanical process. Final heat treatment operations are usually used for stabilization of microstructure (they restrict grain growth) (Motyka & Sieniawski, 2010).

Titanium alloys together with aluminium alloys belong to the largest group of superplastic materials used in industrial SPF. Their main advantages are good superplasticity combined with relatively high susceptibility to diffusion bonding. Among them two-phase $\alpha+\beta$ Ti-6Al-4V alloy has been the most popular for many years as it exhibits superplasticity even after application of conventional plastic working methods (Sieniawski & Motyka, 2007).

2. Deformation behaviour of titanium and its alloys

Titanium has two allotropic forms: Ti_{α} with hexagonal close packed (hcp) crystal structure (up to 882.5°C) and Ti_{β} with body centered cubic (bcc) crystal structure (between 882.5 and 1662°C). Each of the allotropes exhibit different plasticity resulting from its crystal structure and different number of the slip systems (A. D. Mc Quillan & M. K. Mc Quillan, 1956; Zwicker, 1974).

Deformation of α titanium, both at room and elevated temperature, occurs by slip and twinning (Tab. 1). Primary slip systems in α titanium are: $\{1\bar{1}00\} \langle 11\bar{2}0 \rangle$ and $\{1\bar{1}01\} \langle 11\bar{2}0 \rangle$ (Fig. 1). Critical resolved shear stress value (τ_{crss}) depends on deformation temperature, impurities content and slip system (Fig. 2). If the relative value of τ_{crss} stress for $\{1\bar{1}00\} \langle 11\bar{2}0 \rangle$ slip system is set to 1, for $\{1\bar{1}01\} \langle 11\bar{2}0 \rangle$ system it equals to 1.75, and for $\{0001\} \langle 11\bar{2}0 \rangle$ system - 1.92. Coarse grained and single crystal α titanium deforms in $\{0001\} \langle 11\bar{2}0 \rangle$ system, because stacking fault energy and atom packing density are larger in $\{1\bar{1}00\}$ planes than in $\{0001\}$ planes (Kajbyszew & Krajuchijn, 1967; Zwicker, 1974).

Metal	c/a ratio	Slip plane	Critical resolved shear stress τ_{crss} , MPa	Twinning plane
Zinc	1.856	$\{0001\}$ $\{1\bar{1}00\}$	0.34 10-15	$\{1\bar{1}02\}$
Manganese	1.624	$\{0001\}$ $\{1\bar{1}00\}$	4.5 5.1	$\{1\bar{1}02\}$ $\{1\bar{1}01\}$
		$\{1\bar{1}00\}$	39.2	$\{1\bar{1}03\}$ $\{1\bar{1}21\}$
Titanium α	1.587	$\{0001\}$	62.1	$\{1\bar{1}22\}$ $\{1\bar{1}21\}$
		$\{1\bar{1}00\}$	13.83	$\{1\bar{1}22\}$ $\{1\bar{1}02\}$

Table 1. Slip and twinning planes and critical resolved shear stress in metals with hcp crystal structure (Kajbyszew & Krajuchijn, 1967)

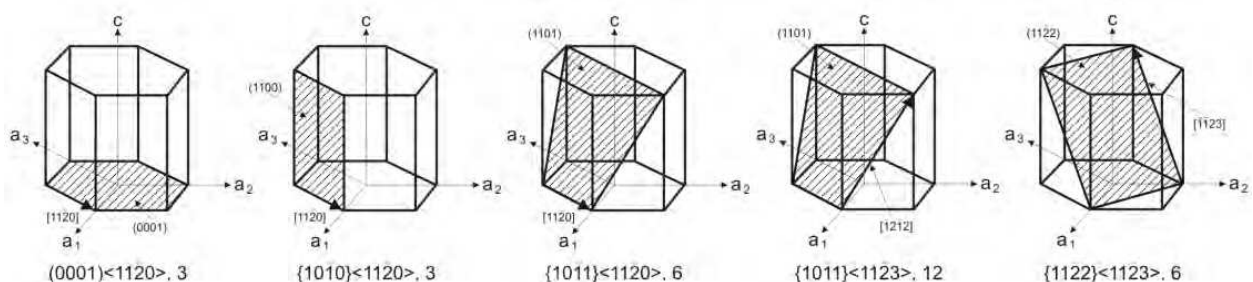


Fig. 1. Slip systems in α titanium (Balasubramanian & Anand, 2002)

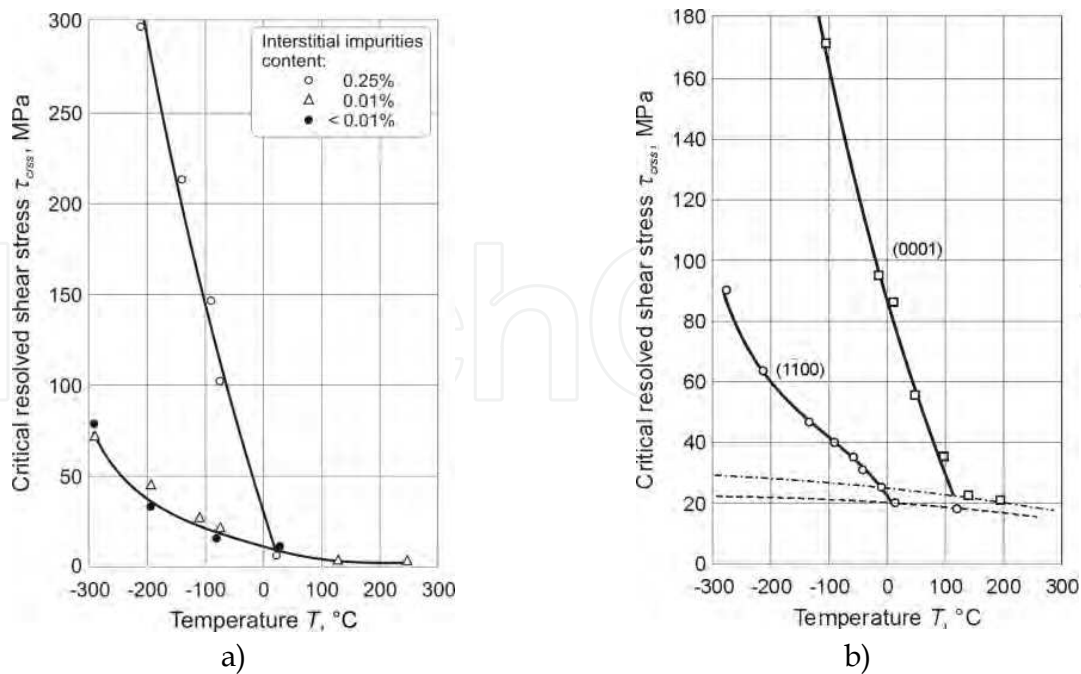


Fig. 2. Critical resolved shear stress as a function of: a) interstitial impurities content and temperature (Zwicker, 1974) b) temperature and slip plane in titanium with 0.1% interstitial impurities content (Levine, 1966)

The number of twinning planes is higher in α titanium than in other metals (except for zirconium) having hcp crystal structure. Twinning in single crystal α titanium occurs in $\{1\bar{1}02\}$, $\{11\bar{2}1\}$, $\{11\bar{2}2\}$, $\{11\bar{2}3\}$, $\{11\bar{2}4\}$ planes, while in polycrystalline α titanium in $\{1\bar{1}02\}$, $\{11\bar{2}1\}$, $\{11\bar{2}2\}$ planes (Fig. 3). The smallest value of critical stress for twinning occurs in $\{1\bar{1}02\}$ plane. It was found that slip and twinning interact with each other. Twinning in $\{11\bar{2}2\}$ plane hinders slip in that plane. Twinning is intensified by increase in deformation, metal purity, grain size and by decrease in temperature (Churchman, 1955).

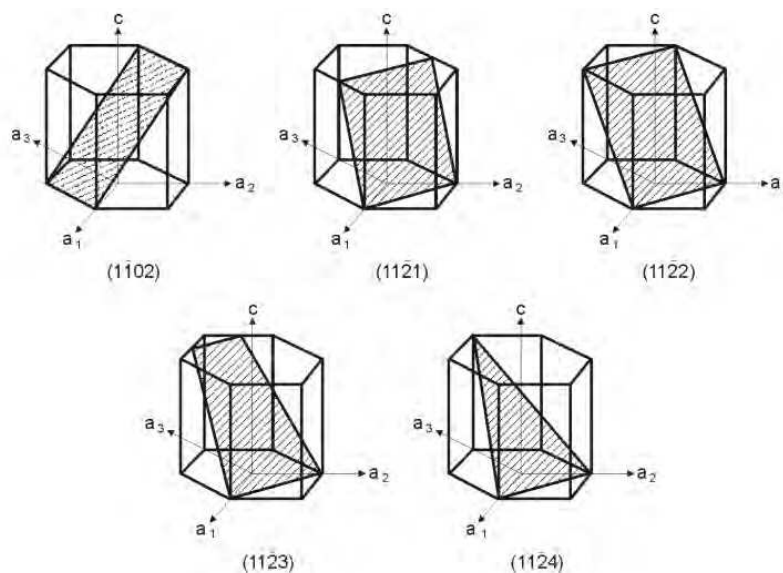


Fig. 3. Twinning planes in α titanium (Balasubramanian & Anand, 2002)

Plastic deformation in β titanium takes place by mechanisms characteristic of metals with bcc structure. In β titanium following slip systems operate: $\{110\} \langle 111 \rangle$, $\{112\} \langle 111 \rangle$, $\{123\} \langle 111 \rangle$ along with twinning system $\{112\} \langle 111 \rangle$ (Balasubramanian & Anand, 2002; Wassermann & Grewen, 1962). Alloying elements affect plasticity of β titanium to different degree. Cold strain of β titanium (with high content of Mo, Nb or Ta) can exceed 90%. However addition of ruthenium or rhodium results in very large decrease in plasticity, rendering cold deformation of β titanium practically impossible (Raub & Röschelb, 1963).

Titanium alloys can be classified according to their microstructure in particular state (e.g. after normalizing). It must be emphasized that this classification is questionable, because phase transformations in alloys with transition elements proceed so slowly, that very often the microstructure consistent with phase equilibrium diagram cannot be obtained at room temperature. According to widely accepted classification of the alloys in normalized state following types of titanium alloys can be distinguished (Fig. 4) (Glazunov & Kolachev, 1980):

1. α alloys
 - not heat strengthened
 - heat strengthened due to metastable phases decomposition
2. $\alpha+\beta$ alloys
 - strengthened by quenching
 - with increased plasticity after quenching
3. β alloys
 - with mechanically unstable β_{MN} phase (decomposing under the stress)
 - with mechanically stable β_{MS} phase (not decomposing under the stress)
 - with thermodynamically stable β_{TS} phase.

Two transition types of alloys can also be distinguished:

1. near α alloys – which contain up to 5% of β stabilizers (microstructure is composed of α phase and small amount, about 3-5%, of β phase),
2. near β alloys – their properties match the properties of $\alpha+\beta$ alloys with high volume fraction of β phase (the microstructure after solutionizing is composed of metastable β_M phase).

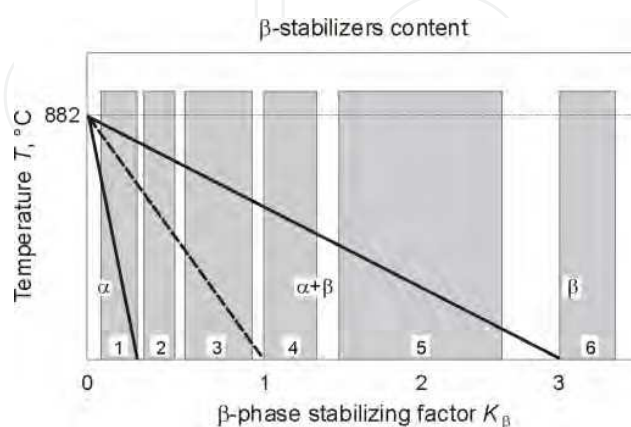


Fig. 4. Classification of titanium alloys on the basis of β stabilizers content: 1 – α alloys, 2 – near α alloys, 3 – martensitic $\alpha+\beta$ alloys, 4 – transition $\alpha+\beta$ alloys, 5 – near β alloys, 6 – β alloys (Glazunov & Kolachev, 1980)

Plasticity of commercial pure (CP) titanium and single-phase alloys depends on type and content of impurities, alloying elements, temperature of deformation and strain rate. Increase in impurities and alloying elements content reduces plasticity due to solid solution strengthening (Fig. 5) (Glazunov & Mojsiejew, 1974).

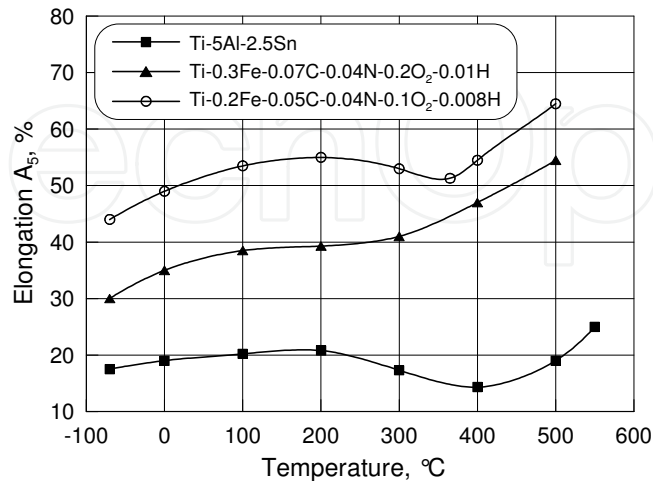


Fig. 5. The influence of temperature on plasticity of CP titanium and single-phase α alloys (Glazunov & Mojsiejew, 1974)

Increase in temperature of deformation reduces plastic flow stress of the CP titanium and single-phase α alloys (Fig. 6).

Critical strain of CP titanium depends on temperature of deformation and strain rate (Fig. 7).

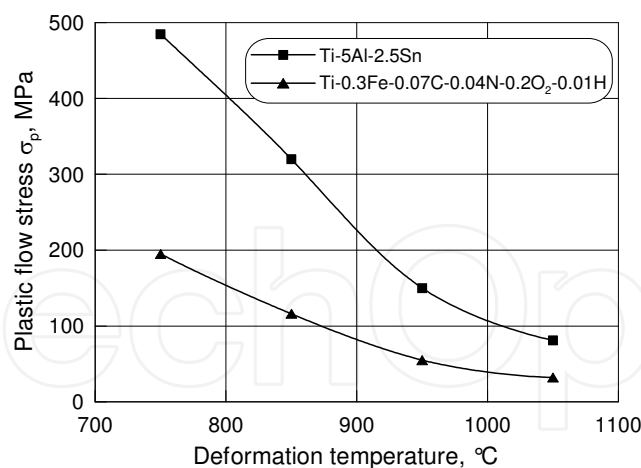


Fig. 6. Dependence of plastic flow stress of CP titanium and α alloys on deformation temperature (Hadasik, 1979)

Roughly 90% of about 70 grades of titanium alloys that are manufactured by conventional methods are two-phase martensitic or transition alloys. They exhibit high relative strength (UTS/ρ), good creep resistance (up to 450°C), corrosion resistance in many environments, good weldability and formability. The most widely used representative of this group of alloys is Ti-6Al-4V showing good balance of mechanical and technological properties.

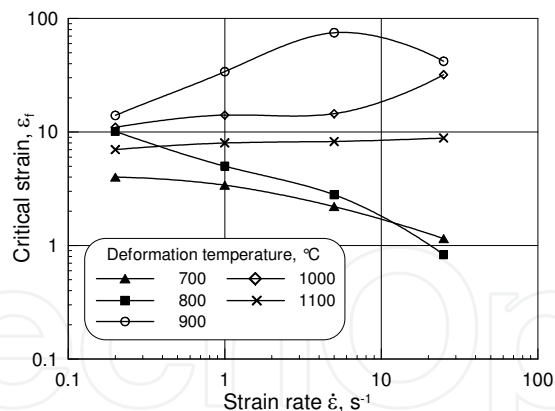


Fig. 7. The effect of temperature of deformation and strain rate on the critical strain of CP titanium (Hadasik, 1979)

The basic technological processes enabling final product manufacturing and development of mechanical properties of titanium alloys are hot working and heat treatment. Application of cold working is limited to the operations of bending (sheets, flat bars, tubes and bars) and shallow drawing of sheets allowing to obtain large elements. Bulk cold forming is not used due to high resistance of titanium and its alloys to plastic flow. (Fig. 8) (Lee & Lin, 1997).

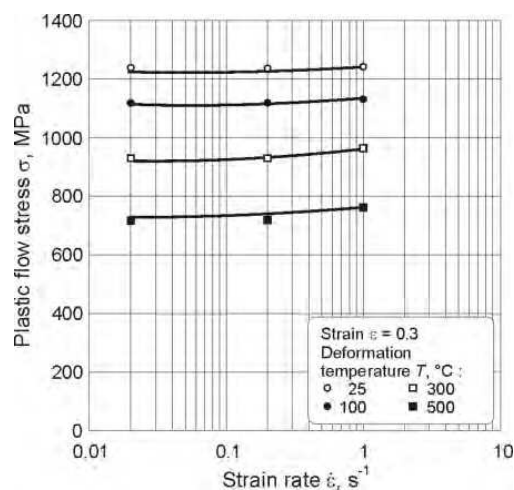


Fig. 8. The influence of strain rate on plastic flow stress of cold worked Ti-6Al-4V alloy (Lee & Lin, 1997)

Plasticity, microstructure and mechanical properties of titanium alloys depend on hot working conditions (Peters et al., 1983; Brooks, 1996):

- heating rate, time and temperature of soaking and furnace atmosphere,
- start and finish temperature of deformation,
- draft in final operations of deformation,
- strain rate,
- cooling rate after deformation.

Application of all or selected operations of hot working and heat treatment (Fig. 9) allows to vary to a large extent the microstructure e.g. morphology and dispersion of phases in $\alpha+\beta$ titanium alloys (Fig. 10).

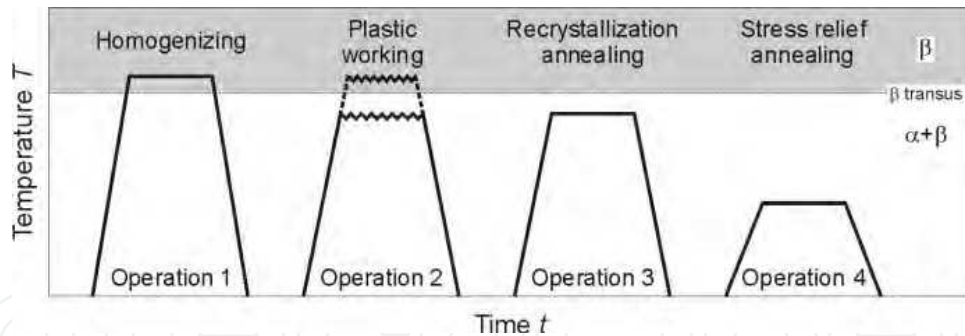


Fig. 9. The schematic diagram of plastic work and heat treatment processes of two phase titanium alloys (Peters et al., 1983)

The processes of hot working and heat treatment of two-phase titanium alloys, e.g. Ti-6Al-4V, allow to obtain various types of microstructure (Fig.10) (Ezugwu & Wang, 1997; Kubiak & Sieniawski, 1998):

- martensitic (or composed of metastable β_M phase),
- globular (fine or coarse-grained),
- necklase (fine or coarse-grained),
- lamellar (fine or coarse-grained),
- bi-modal (with various volume fraction and dispersion of α phase).

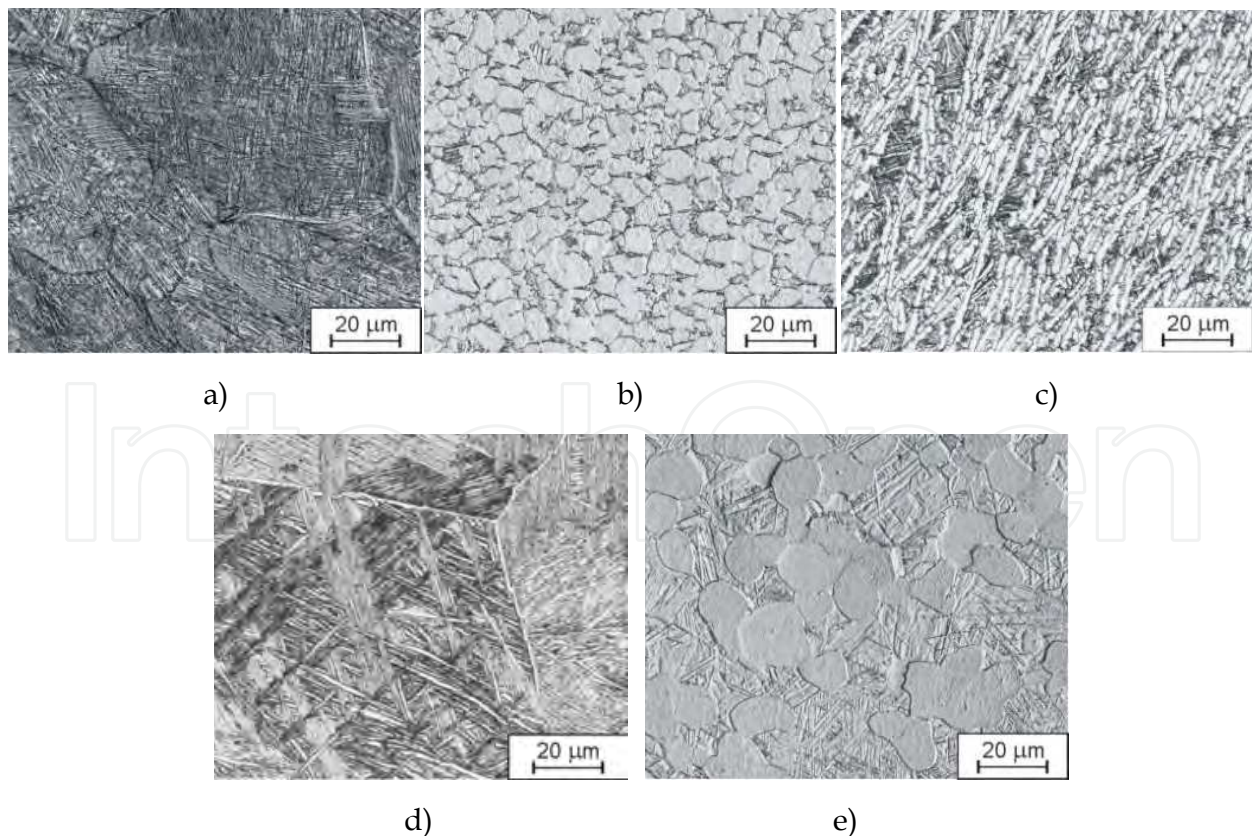


Fig. 10. Microstructure of two-phase Ti-6Al-4V alloy: a) martensitic, b) globular, c) necklase, d) lamellar, e) bi-modal (Kubiak & Sieniawski, 1998)

3. The influence of deformation conditions and morphology of phases on the plasticity of $\alpha+\beta$ titanium alloys

Hot deformation behaviour of two-phase titanium alloys depends on chemical and phase composition, stereological parameters of microstructure and process conditions (deformation temperature, strain rate, stress and strain distribution). They exhibit analogous dependences of the plastic flow stress on temperature and strain rate like other metals alloys. Increase in strain rate raises plastic flow stress while increase in deformation temperature reduces it (Sakai, 1995).

Increase in temperature of deformation of two-phase titanium alloys reduces plastic flow stress (Fig. 11) more effectively in $\alpha+\beta$ field than in β phase field, as a result of change in volume fraction of α and β phases (Fig. 12) (Ding et al., 2002; Sheppard & Norley, 1988).

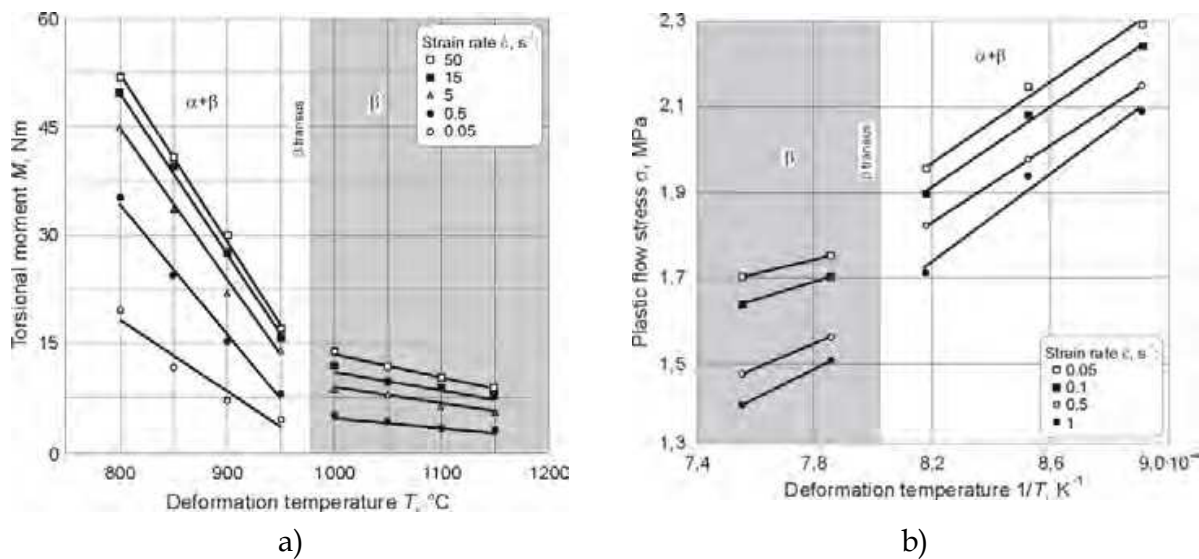


Fig. 11. The effect of temperature of deformation and strain rate on plastic flow stress for Ti-6Al-4V alloy: a) torsion test (Sheppard & Norley, 1988), b) compression test (Ding et al., 2002)

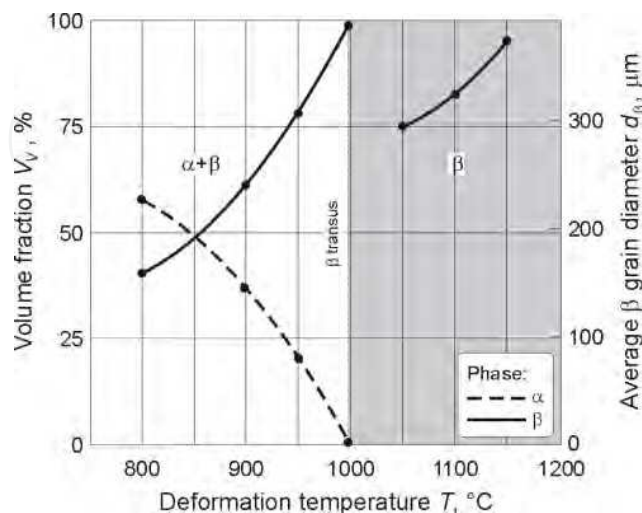


Fig. 12. The effect of temperature on volume fraction of α and β phases in Ti-6Al-4V alloy during torsion test (Sheppard & Norley, 1988)

High sensitivity to strain rate is a characteristic feature of two-phase titanium alloys. The coefficient of strain rate sensitivity depends on temperature of deformation and strain rate (Fig. 13), grain size (Fig. 14), strain magnitude (Fig. 15) and morphology of phase constituents (Fig. 16) (Semiatin et al., 1998).

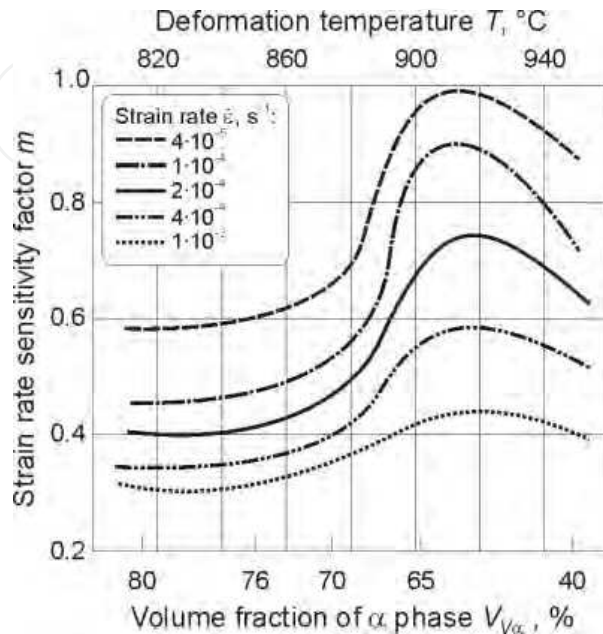


Fig. 13. The dependence of strain rate sensitivity factor m on temperature (volume fraction of α phase) and strain rate for Ti-6Al-4V alloy (Semiatin et al., 1998)

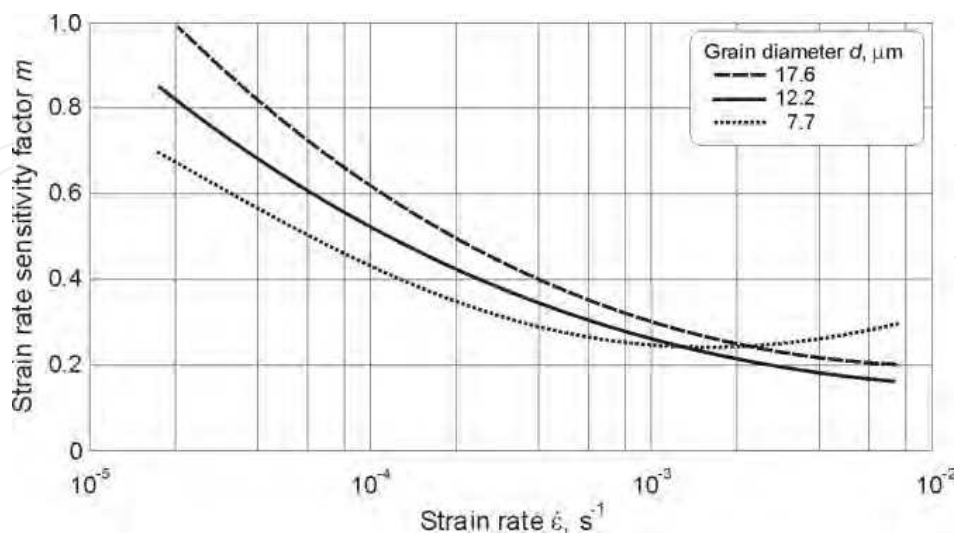


Fig. 14. The dependence of strain rate sensitivity factor m on grain size of α phase for Ti-6Al-4V alloy (Semiatin et al., 1998)

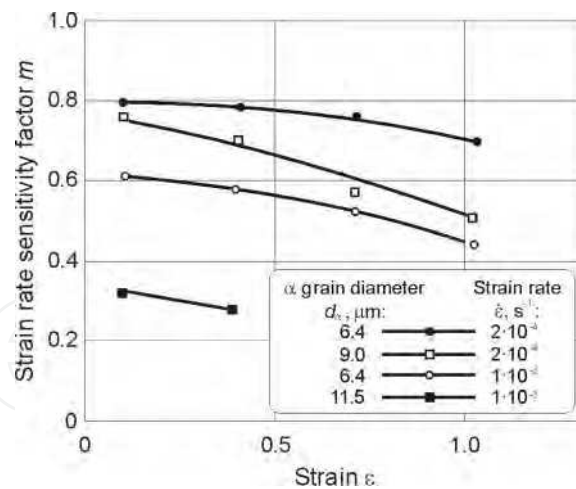


Fig. 15. The dependence of strain rate sensitivity factor m on the strain magnitude for Ti-6Al-4V alloy (Ghosh & Hamilton, 1979)

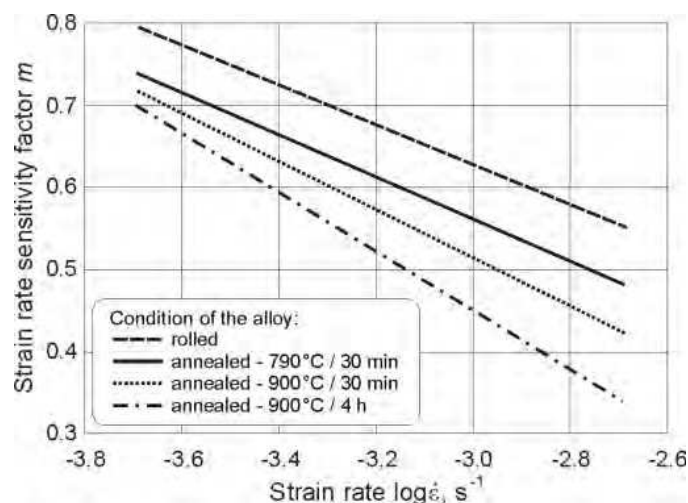


Fig. 16. The dependence of strain rate sensitivity factor m on microstructure morphology (type of heat treatment) for hot-rolled sheets of Ti-6Al-4V alloy (Semiatin et al., 1998)

Volume fraction of α and β phases affects the value of the coefficient of strain rate sensitivity m . For two-phase titanium alloys it reaches maximum value when $V_{v\alpha} = V_{v\beta} \approx 50\%$, at the temperature close to start temperature of $\alpha + \beta \rightarrow \beta$ phase transformation and strain rate in the range of $\dot{\epsilon} = 4 \cdot 10^{-5} - 1 \cdot 10^{-3} \text{ s}^{-1}$. Grain refinement leads to increase in m value and growth of strain magnitude reduces it (Ghosh & Hamilton, 1979).

Deformation of titanium alloys in two-phase range leads to distortion of α and β grains, fragmentation of α grains and their globularization. As a result of these processes (Fig. 17) elongated α phase grains develop with particular orientation which are arranged along direction of maximum deformation (Seshacharyulu et al., 2002).

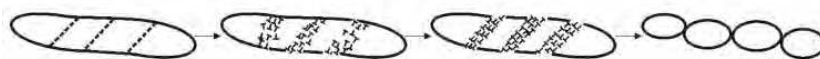


Fig. 17. The deformation process in $\alpha + \beta$ field - fragmentation and globularization of α phase lamellae (Seshacharyulu et al., 2002)

These processes occur simultaneously during deformation and have an impact both on texture (Fig. 18) and morphology of phases (Lütjering, 1998).

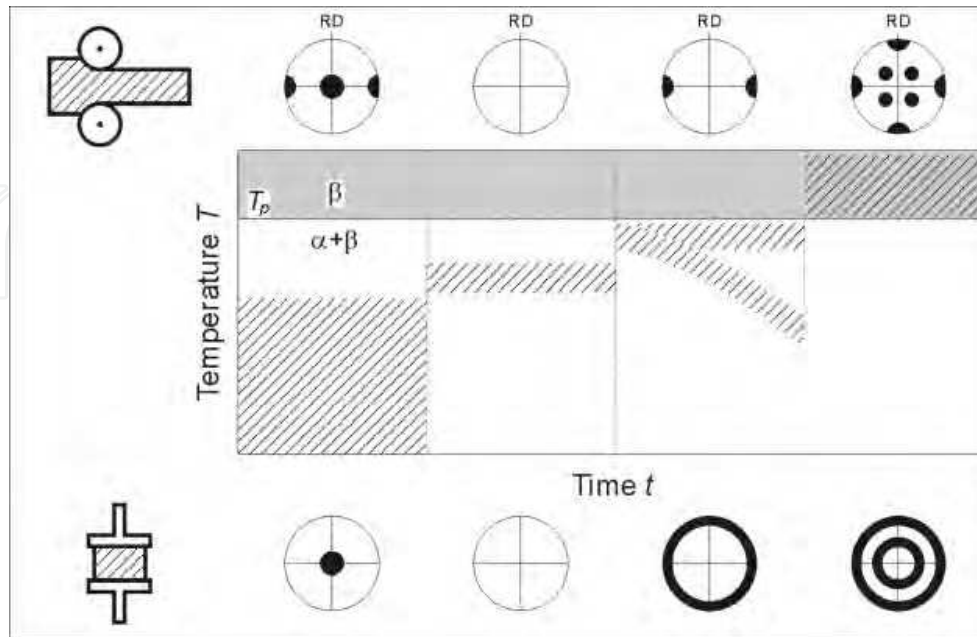


Fig. 18. The texture diagram – pole figure (0002) – of two-phase titanium alloys for various modes and temperature of deformation (Lütjering, 1998)

After deformation in the $\alpha+\beta$ temperature range and air cooling following microstructure of α phase can be obtained (Fig. 19) (Kubiak & Sieniawski, 1998):

- coarse lamellar – not deformed or slightly deformed large size lamellae of primary α phase are usually present in dead zone of semi-finished products (Figs 19-3 and 19-5),
- fine lamellar – α lamellae formed upon cooling from temperature above the β transus temperature at the cooling rate slightly lower than critical (Fig. 19-1),
- distorted lamellar – primary α lamellae which were fragmented and deformed (Fig. 19-4),
- equiaxed – α grains formed upon slow cooling from the temperature higher than finish temperature of $\alpha+\beta\leftrightarrow\beta$ transformation (Fig. 19-2),
- distorted equiaxed – α grains distorted upon deformation below the finish temperature of $\alpha+\beta\leftrightarrow\beta$ transformation (Fig. 19-6).

Increase in degree or rate of deformation may lead to local increase in temperature even above finish temperature of $\alpha+\beta\leftrightarrow\beta$ transformation and development of martensitic α' (α'') phases or colonies of lamellar α and β phases with stereological parameters depending on cooling rate.

Upon deformation of $\alpha+\beta$ titanium alloys above the finish temperature of $\alpha+\beta\leftrightarrow\beta$ transformation dynamic recrystallization occurs. New grains nucleate at primary β grain boundaries and then grow until complete ring arranged along these grain boundaries is formed. New nuclei form also inside formerly recrystallized grains in the ring and the process repeats until new grains completely fill up the 'old' grain (Fig. 20) (Ding et al., 2002).

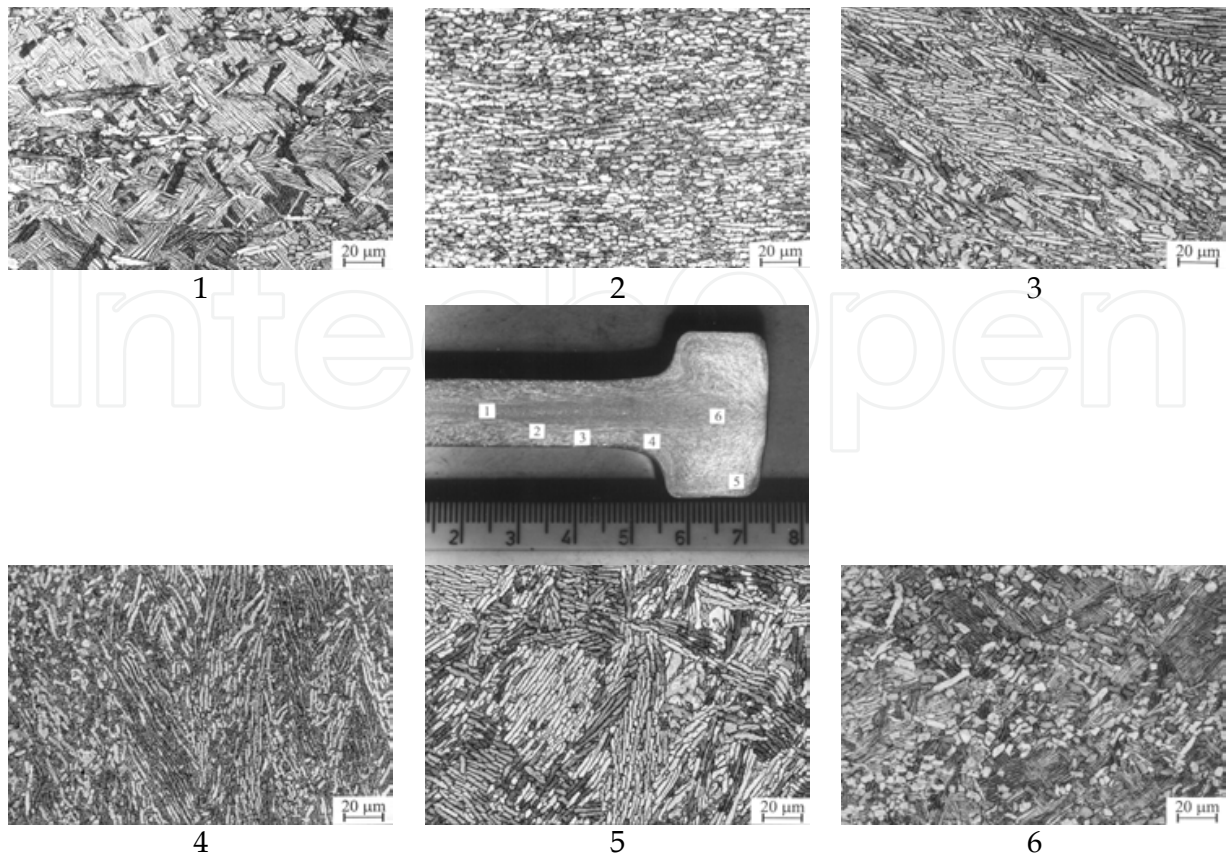


Fig. 19. The microstructure of Ti-6Al-4V alloy after die forging at 950°C (Kubiak & Sieniawski, 1998)

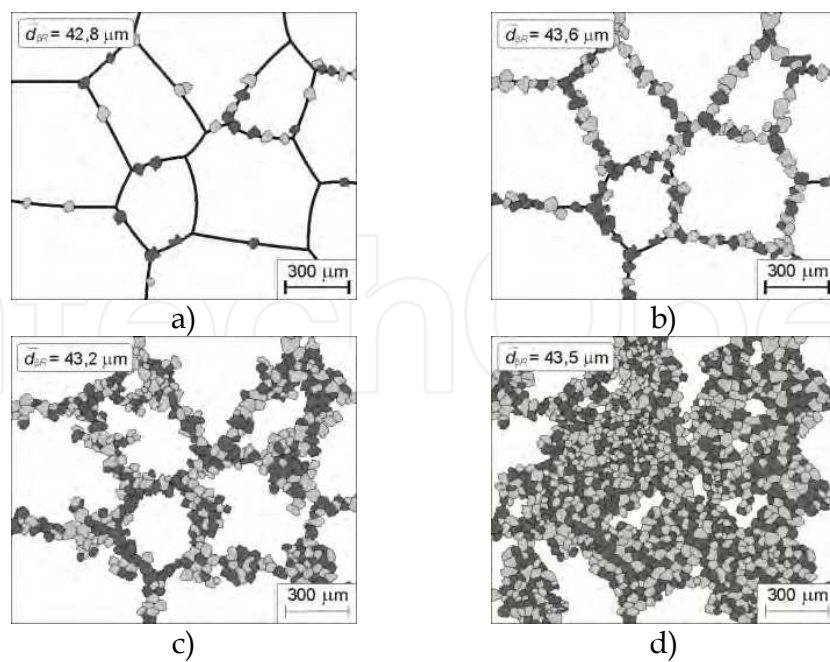


Fig. 20. Simulation of microstructure development for Ti-6Al-4V alloy deformed at 1050°C at the strain rate $\dot{\epsilon} = 1.0 \text{ s}^{-1}$ for various strains: a) $\epsilon = 0.7$; b) $\epsilon = 5.0$; c) $\epsilon = 15$, d) $\epsilon = 45$; $\bar{d}_{\beta R}$ - average diameter of recrystallized β grains (Ding et al., 2002)

Increase in deformation rate decreases the size of recrystallized α grains (Fig. 21). The change in average diameter of primary β grains - $\bar{d}_{p\beta}$ and average thickness of α lamellae - $\bar{g}_{p\alpha}$, induced by deformation, can be described for Ti-6Al-4V alloy by following equations, depending on morphology of primary α and β phases:

- lamellar microstructure $\bar{d}_{p\beta} = 1954.3 Z^{-0.172}$ (μm) (Seshacharyulu et al., 2002)
- lamellar microstructure $\bar{g}_{p\alpha} = 1406.4 Z^{-0.139}$ (μm) (Seshacharyulu et al., 2002)
- equiaxed microstructure $\log(\bar{d}_{p\beta}) = 3.22 - 0.16 \log(Z)$ (Seshacharyulu et al., 1999)

where: Z - Zener-Holomon parameter.

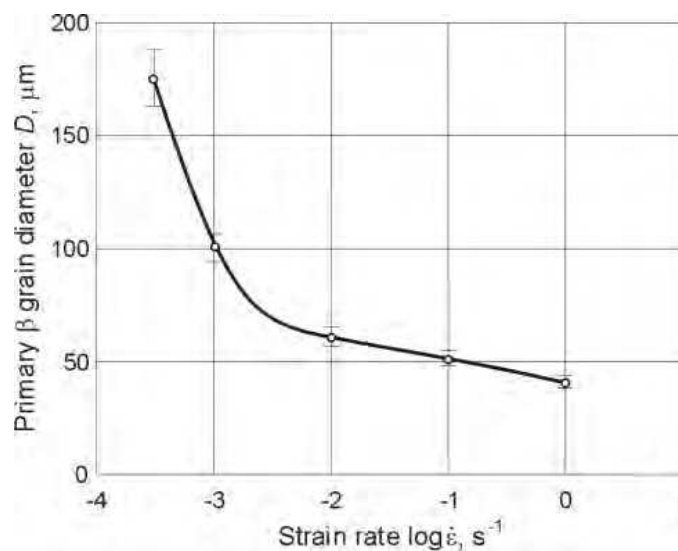


Fig. 21. The effect of strain rate on average diameter of primary β phase grains in Ti 6Al-4V alloy with globular initial microstructure (Seshacharyulu et al., 1999)

However presented model of dynamic recrystallization (Fig. 20) does not take into account the change of shape of primary β grains and deformation of recrystallized β grains. Thus it can be stated that it describes process of metadynamic recrystallization. Upon cooling after dynamic or metadynamic recrystallization following phases can be formed within β grains (depending on cooling rate): colonies of α and β lamellae, equiaxed α and β grains, martensitic α' (α'') phases or the mixture of them.

The character of flow curves obtained during sequential deformation of two phase titanium alloys at the temperature in β field confirms occurrence of dynamic processes of microstructure recovery. Increase in time of metadynamic recrystallization reduces strengthening. The character of the curves describing volume fraction of recrystallized β phase confirms the influence of chemical composition on recrystallization kinetics (Fig. 22). Increasing content of alloying elements leads to decrease in rate of recrystallization. Time for recrystallization of 50% of β phase is equal $t_{0,5} \approx 3.5$ s for Ti-6Al-5Mo-5V-2Cr-1Fe alloy and $t_{0,5} \approx 1.5$ s for Ti-6Al-4V alloy.

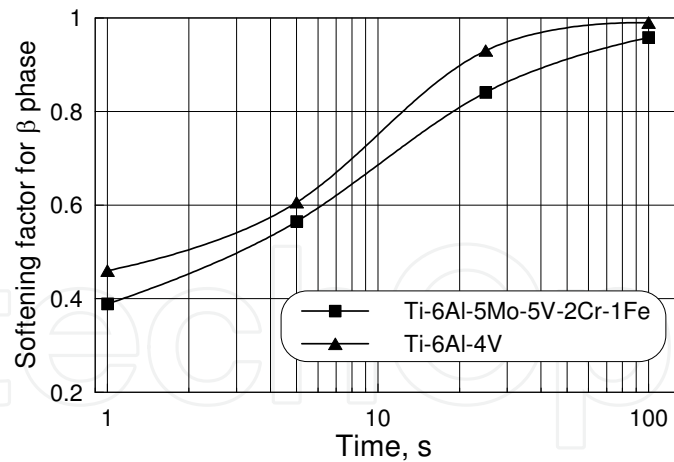


Fig. 22. The dependence of volume fraction of recrystallized β phase on dwell time during sequential deformation of Ti-6Al-4V and Ti-6Al-5V-5Mo-2Cr-1Fe alloys (Kubiak, 2004)

Microstructure of $\alpha+\beta$ titanium alloys after sequential deformation at the temperature in β field and air cooling is composed of globular and heavily deformed α grains (Fig. 23) in the matrix of the lamellar α and β phases (Fig. 24a, b). Increase in dwell time during sequential deformation leads to reduction of the strengthening effect and moreover to reduction of dispersion of the phases – growth of both globular and deformed α grains. New, recrystallized grains nucleate at the primary β grain boundaries, forming chains (Figs 23b, 24b). The dislocation density in globular and lamellar α phase and β phase is low (Fig. 24c, d).

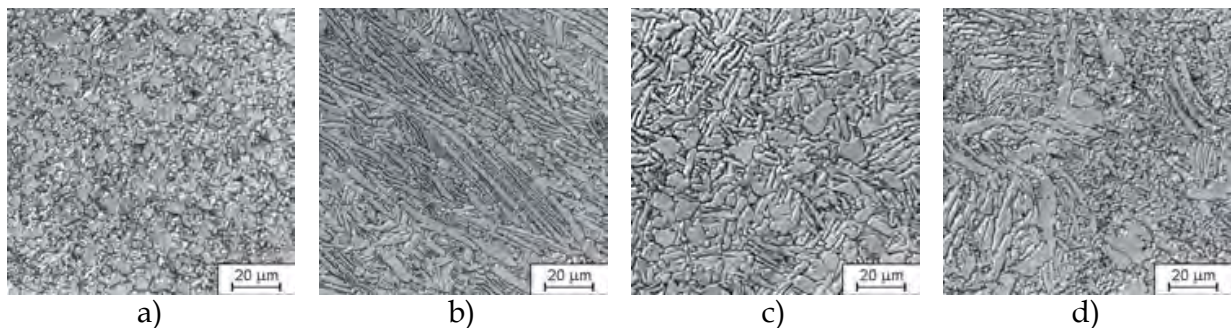


Fig. 23. The microstructure of Ti-6Al-4V alloy after sequential deformation with various dwell time: a) 1s – transverse section, b) 1s – longitudinal section, c) 100s – transverse section, b) 100s – longitudinal section (Kubiak, 2004)

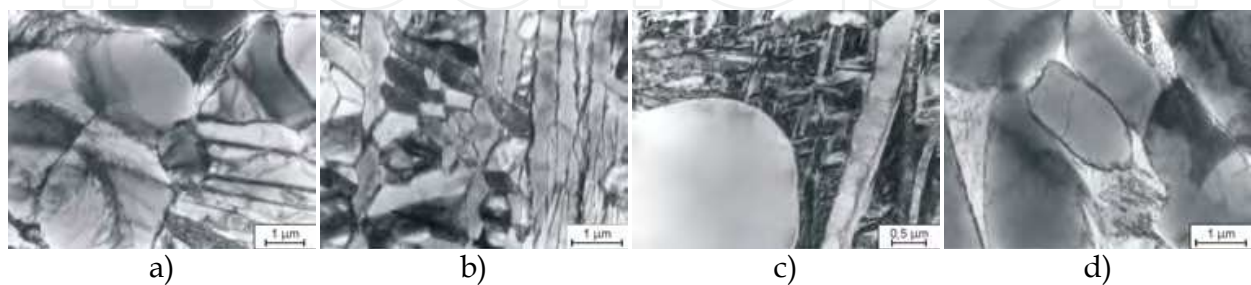


Fig. 24. The microstructure of Ti-6Al-4V alloy after sequential deformation with various dwell time: a) 1s – transverse section, b) 1s – longitudinal section, c) 100s – transverse section, b) 100s – longitudinal section (Kubiak, 2004)

In the 1980s and 1990s deformation maps for CP titanium were calculated on the basis of material constants, describing possible deformation mechanisms operating during processing. In the late 1990s the map of microstructure changes depending on deformation conditions was developed for Ti-6Al-4V alloy (Fig. 25) (Seshacharyulu et al., 1999).

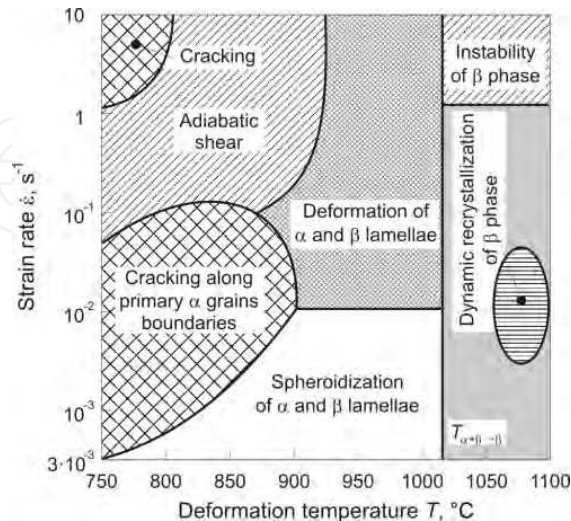


Fig. 25. The map of microstructure changes and phenomena occurring during hot deformation of Ti-6Al-4V alloy (Seshacharyulu et al., 1999)

Technological hot plasticity of $\alpha+\beta$ titanium alloys, characterized by plastic flow stress and critical strain, depends on morphology and stereological parameters of the phases in the alloy microstructure and deformation conditions (Kubiak, 2004).

The flow curves $\sigma = f(\epsilon)$ of $\alpha+\beta$ titanium alloys have similar character. Three stages of flow stress changes can be distinguished, what is characteristic for materials in which dynamic recrystallization occurs (Fig. 26):

- increase up to σ_{pm} value,
- decrease down to σ_{ps} value,
- stabilization at σ_{ps} value.

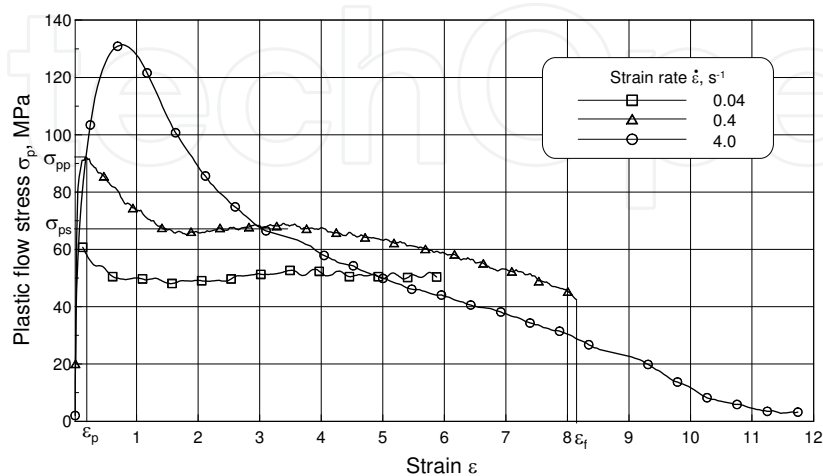


Fig. 26. The effect of strain rate on plastic flow stress at 900°C for Ti-6Al-4V alloy - coarse-grained lamellar microstructure (Kubiak, 2004)

Deformation of Ti-6Al-4V alloy at 1050°C – the range of β phase stability – results in reduction of maximum plastic flow stress σ_{pm} regardless of the initial phase morphology and dispersion. Coarse lamellar microstructure shows the maximum plastic flow stress. Reduction of the size of colonies and lamellae of α and β phases leads to significant decrease in flow stress in comparison with bi-modal and globular microstructure. It was also found that deformation rate have a pronounced effect on the flow stress σ_{pm} and ϵ_m strain for Ti-6Al-4V alloy. Increase in strain rate results in higher σ_{pm} and ϵ_m values (Figs 27 and 28).

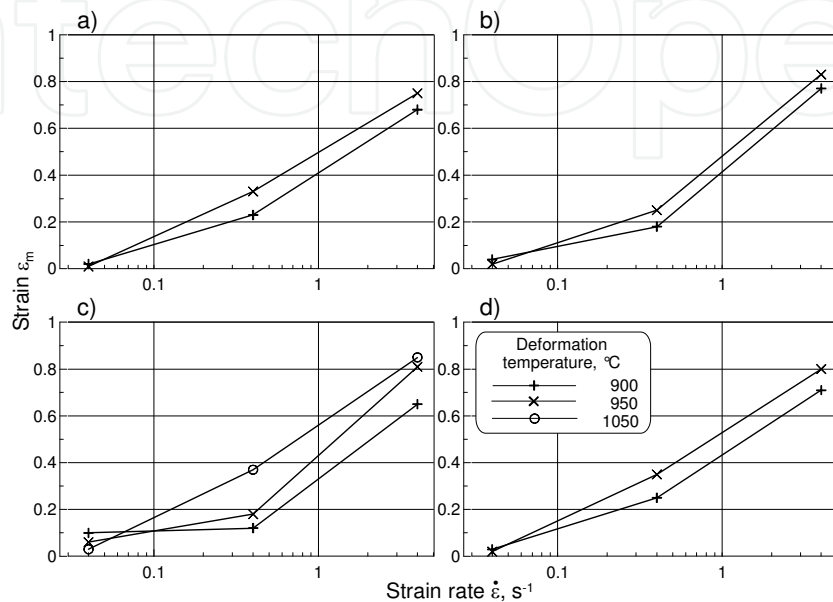


Fig. 27. The dependence of ϵ_m strain on strain rate, temperature of deformation and microstructure of Ti-6Al-4V alloy: a) bi-modal, b) globular, c) coarse-grained lamellar, d) fine-grained lamellar microstructure (Kubiak, 2004)

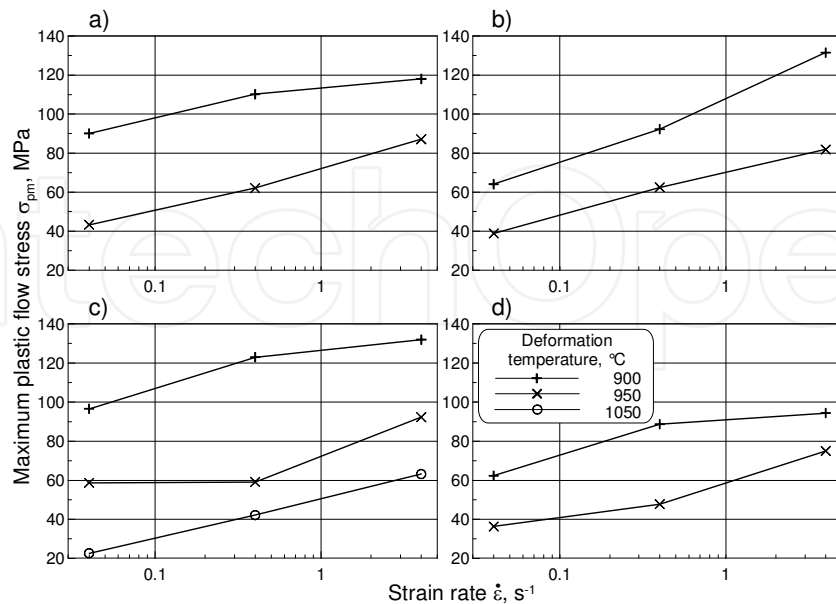


Fig. 28. The dependence of maximum plastic flow stress σ_{pm} on strain rate, temperature of deformation and microstructure of Ti-6Al-4V alloy: a) bi-modal, b) globular, c) coarse-grained lamellar, d) fine-grained lamellar microstructure (Kubiak, 2004)

The influence of microstructure morphology and deformation condition on maximum flow stress σ_{pm} for Ti-6Al-5Mo-5V-2Cr-1Fe alloy is similar to that found for Ti-6Al-4V alloy.

The effect of conditions of heat treatment and degree of plastic deformation in thermomechanical process on development of microstructure and plasticity of Ti-6Al-4V and Ti-6Al-2Mo-2Cr titanium alloys in hot tensile test was also investigated (Motyka & Sieniawski, 2010). On the basis of dilatometric results and previous findings conditions of heat treatment and plastic deformation were defined and two schemes of thermomechanical processing were worked out, denoted TMP-I and TMP-II respectively (Fig. 29).

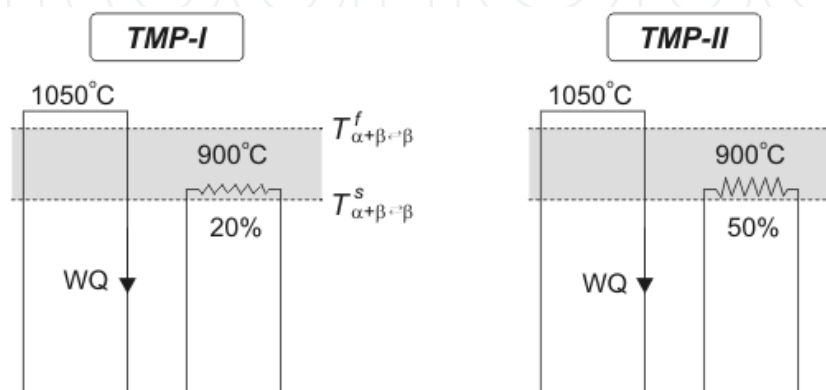


Fig. 29. Schemes of thermomechanical processing of Ti-6Al-4V alloy with forging reduction $\varepsilon \approx 20\%$ (a) and $\varepsilon \approx 50\%$ (b) (WQ - water quenching) (Motyka & Sieniawski, 2010)

Initial microstructure of Ti-6Al-4V alloy was composed of globular, fine α grains and β phase in the form of thin layers separating α grains (Fig. 30a). Quenching of Ti-6Al-4V alloy from the β phase temperature range led to formation of microstructure composed solely of martensitic α' (α'') phase (Fig. 30b). Microstructure after following plastic deformation in the $\alpha+\beta\leftrightarrow\beta$ range with forging reduction of about 20% (TMP-I) and 50% (TMP-II) comprised elongated and deformed grains of primary α phase in the matrix of β transformed phase containing fine globular grains of α secondary phase (Fig. 30c,d). Higher degree of initial deformation led to obtaining finer microstructure containing more elongated α grains - $f_\alpha = 16$ for $\varepsilon = 20\%$ and 21.1 for $\varepsilon = 50\%$. The larger volume fraction of α phase was also found (Tab. 2).

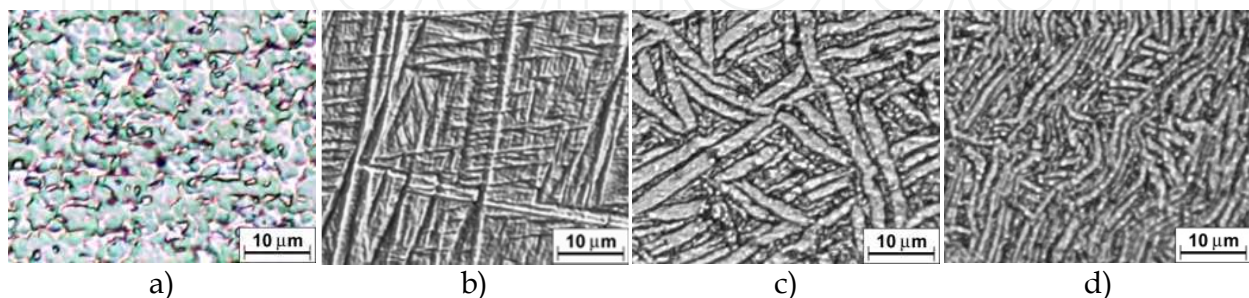


Fig. 30. Microstructure (DIC) of Ti-6Al-4V alloy before thermomechanical processing (a), after quenching from the β phase range (b) and after deformation in the $\alpha+\beta\leftrightarrow\beta$ range with forging reduction of 20% (c) and 50% (d) (Motyka & Sieniawski, 2010)

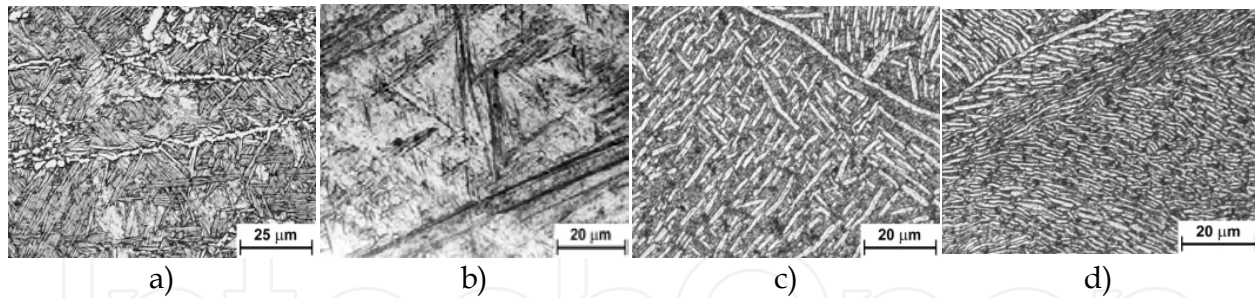


Fig. 31. Microstructure (DIC) of Ti-6Al-2Mo-2Cr alloy before thermomechanical processing (a), after quenching from the β phase range (b) and after deformation in the $\alpha+\beta\leftrightarrow\beta$ range with forging reduction of 20% (c) and 50% (d) (Motyka & Sieniawski, 2010)

Initial microstructure of Ti-6Al-2Mo-2Cr alloy was composed of colonies of parallel α -lamellae enclosed in primary β phase grains (Fig. 31a). Solution heat treatment led to formation of microstructure composed of martensitic α' (α'') phase, similarly to Ti-6Al-4V alloy (Fig. 31b). Microstructure after thermomechanical processes (TMP-I and TMP-II) comprised fine, elongated grains of α phase in the matrix of β transformed phase (Figs 31c,d). In contrary to Ti-6Al-4V alloy primary β phase grain boundaries were observed. Higher degree of initial deformation in thermomechanical process led to obtaining finer microstructure and larger volume fraction of α phase (Table 2).

Condition of Ti-6Al-4V alloy	Stereological parameters					
	V_α %	\bar{a}_α	\bar{b}_α	\bar{f}_α		
		μm				
As received	82	4.1	5.3	0.77		
TMP-I processed	59	51.3	3.2	16		
TMP-II processed	79	23.2	1.1	21.1		
Condition of Ti-6Al-2Mo-2Cr alloy	V_α %	$\bar{a}_{\beta prim}$	$\bar{b}_{\beta prim}$	$\bar{f}_{\beta prim}$	R	g
		μm			μm	
	As received	76	137	42	3.26	12
TMP-I processed	34	-	-	-	-	4
TMP-II processed	40	-	-	-	-	1

Table 2. Stereological parameters of microstructure of as-received and thermomechanically processed Ti-6Al-4V and Ti-6Al-2Mo-2Cr alloys; where: V_α - volume fraction of α phase, \bar{a}_α and \bar{b}_α - length of sides of rectangular circumscribed on α grain section, \bar{f}_α - elongation factor of α phase grains, $\bar{a}_{\beta prim}$ and $\bar{b}_{\beta prim}$ - length of sides of rectangular circumscribed on primary β grain section, $\bar{f}_{\beta prim}$ - elongation factor of primary β phase grains, R - size of the colony of parallel α lamellae, g - thickness of α -lamellae (Motyka & Sieniawski, 2010)

TEM examination of Ti-6Al-4V alloy revealed fragmentation of elongated α grains (Fig. 32a) and presence of globular secondary α grains in the β transformed matrix (Fig. 32b) after TMP-I thermomechanical processing. Higher dislocation density in elongated α grains was observed after TMP-II processing (larger forging reduction) (Figs 33a,b).

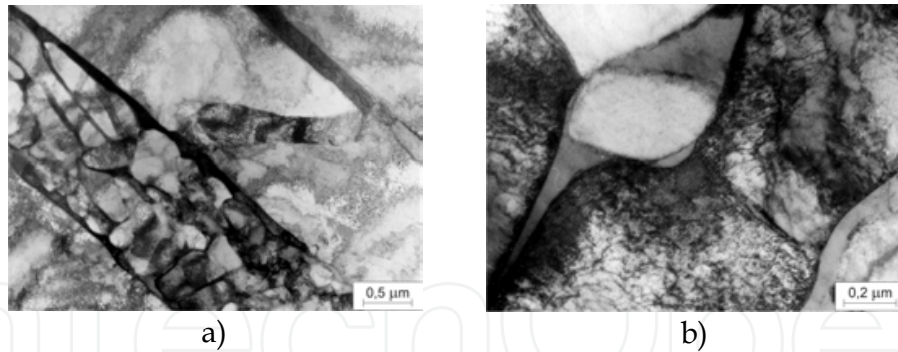


Fig. 32. Microstructure (TEM) of Ti-6Al-4V alloy after TMP-I process: fragmentation of α phase (a), globular secondary α phase grain (b) (Motyka & Sieniawski, 2010)

In Ti-6Al-2Mo-2Cr alloy after TMP-I processing dislocations were observed mainly near grain boundaries (Fig. 34a). It was found that the secondary α phase in β transformed matrix occurs in lamellar form (Fig. 34b). Higher degree of deformation in TMP-II process led to higher dislocation density in α phase grains (Fig. 35a) and fragmentation of elongated α grains (Fig. 35b).

In Ti-6Al-2Mo-2Cr alloy higher volume fraction of β phase (Tab. 3) was found than in Ti-6Al-4V alloy which can be explained by higher value of coefficient of β phase stabilisation K_β .

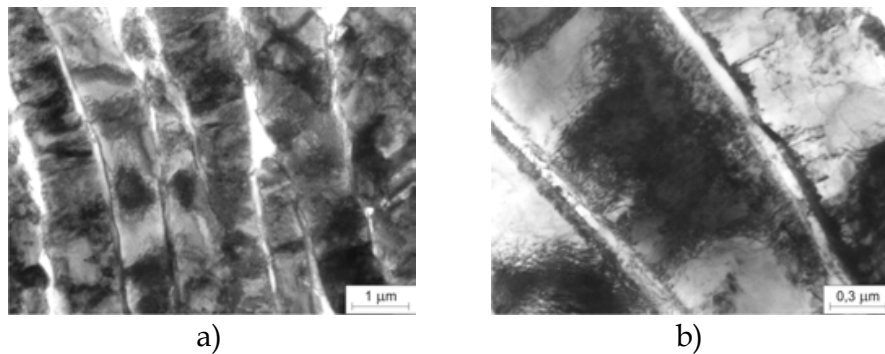


Fig. 33. Microstructure (TEM) of Ti-6Al-4V alloy after TMP II process: high dislocation density in α grains (Motyka & Sieniawski, 2010)

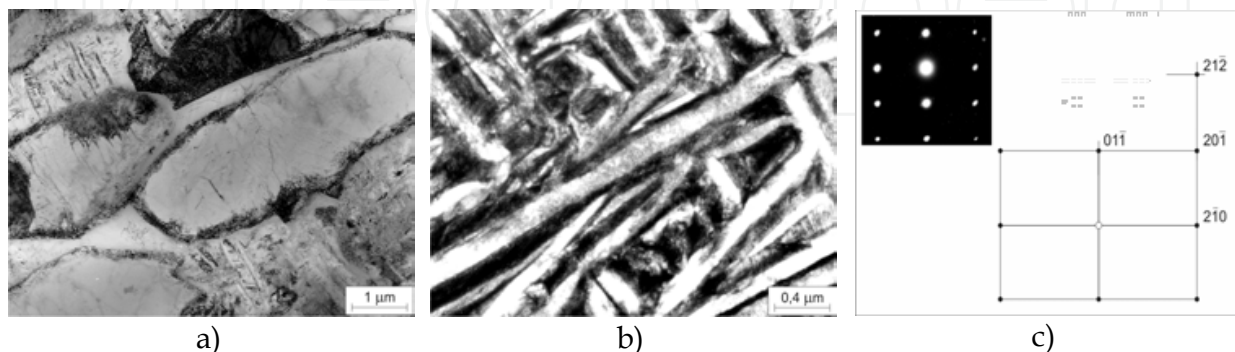


Fig. 34. Microstructure (TEM) of Ti-6Al-2Mo-2Cr alloy after TMP-II process: dislocations in α grains (a), lamellae precipitations of α grains in β transformed phase (b) with indexing of diffraction (c) (Motyka & Sieniawski, 2010)

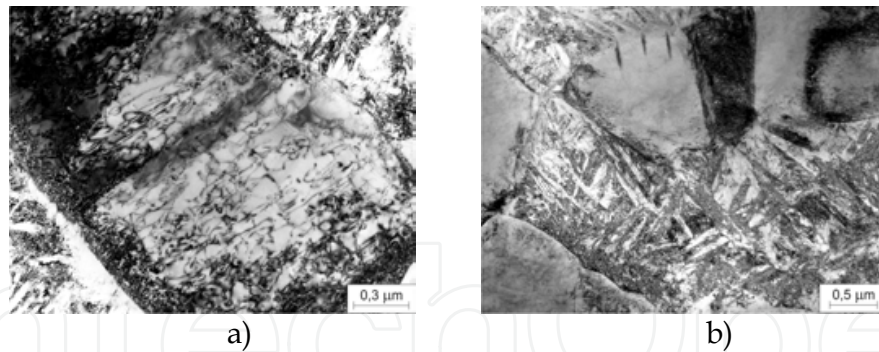


Fig. 35. Microstructure (TEM) of Ti-6Al-2Mo-2Cr alloy after TMP-II process: dislocations in α grains (a), precipitations of α grains in β transformed phase (b) (Motyka & Sieniawski, 2010)

Critical temperatures of $\alpha+\beta\leftrightarrow\beta$ phase transformation [°C]	Condition of Ti-6Al-4V alloy		
	As-received	TMP-I processing	TMP-II processing
Start of $\alpha+\beta\leftrightarrow\beta$	894	882	912
Finish of $\alpha+\beta\leftrightarrow\beta$	979	976	1009
	Condition of Ti-6Al-2Mo-2Cr alloy		
	As-received	TMP-I processing	TMP-II processing
Start of $\alpha+\beta\leftrightarrow\beta$	803	800	809
Finish of $\alpha+\beta\leftrightarrow\beta$	991	992	1011

Table 3. Critical temperatures of $\alpha+\beta\leftrightarrow\beta$ phase transformation of as received and thermomechanically processed two-phase alloys (Motyka & Sieniawski, 2010)

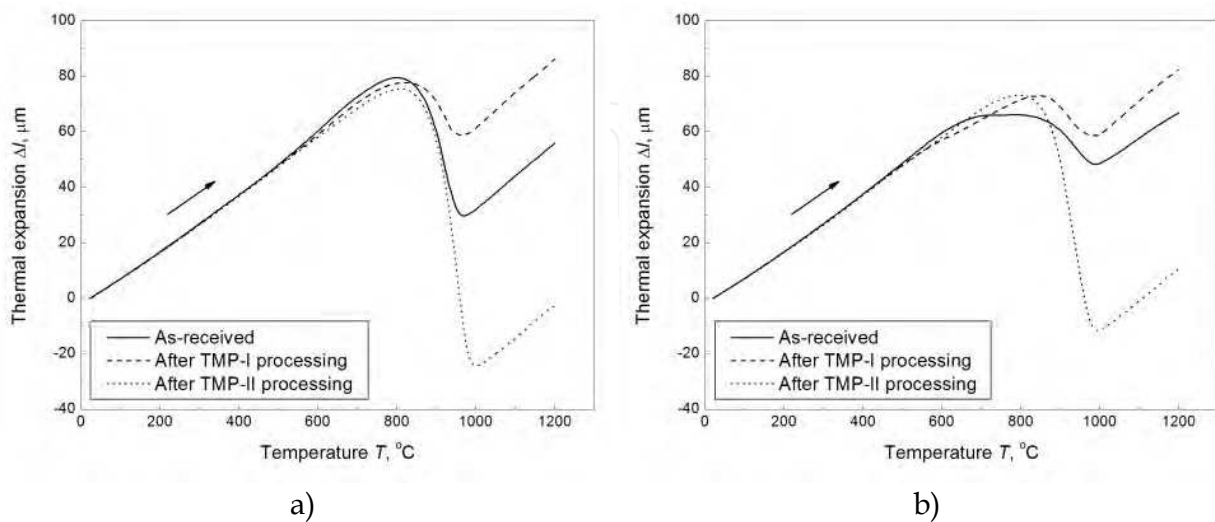
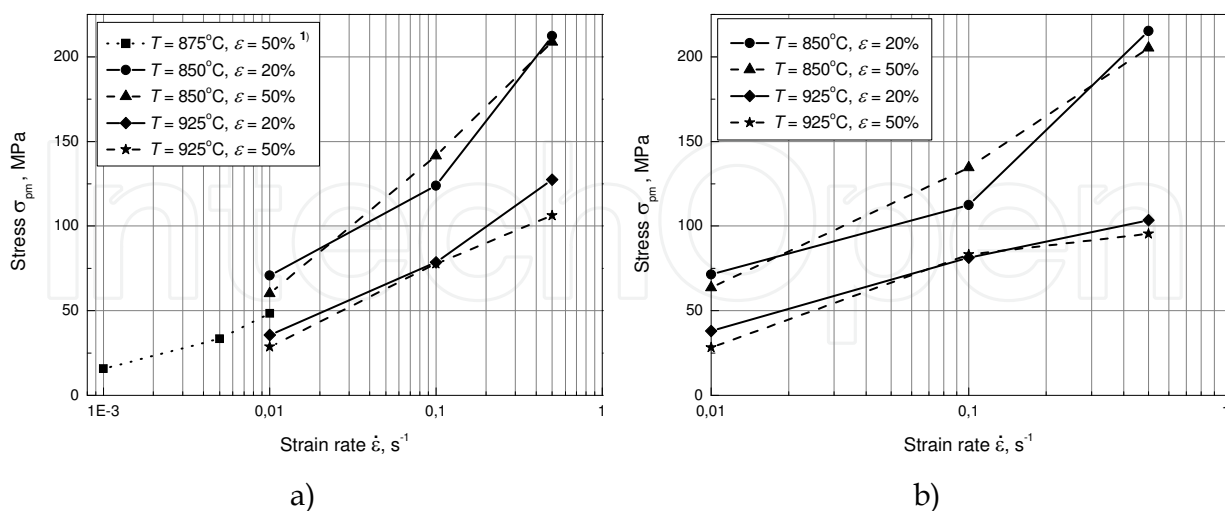


Fig. 36. Dilatometric curves of Ti-6Al-4V (a) and Ti-6Al-2Mo-2Cr (b) alloys in as-received state and after thermomechanical processing (Motyka & Sieniawski, 2010)

Dilatometric examination revealed the influence of forging reduction on critical temperatures of $\alpha+\beta\leftrightarrow\beta$ phase transformation. The TMP-II thermomechanical processing with highest strain applied ($\varepsilon\approx 50\%$) caused significant increase in finish temperature of $\alpha+\beta\leftrightarrow\beta$ phase transformation in both examined titanium alloys (Tab. 3 and Fig. 36). The temperature range of phase transformation was considerably wider in Ti-6Al-2Mo-2Cr alloy (Tab. 3).

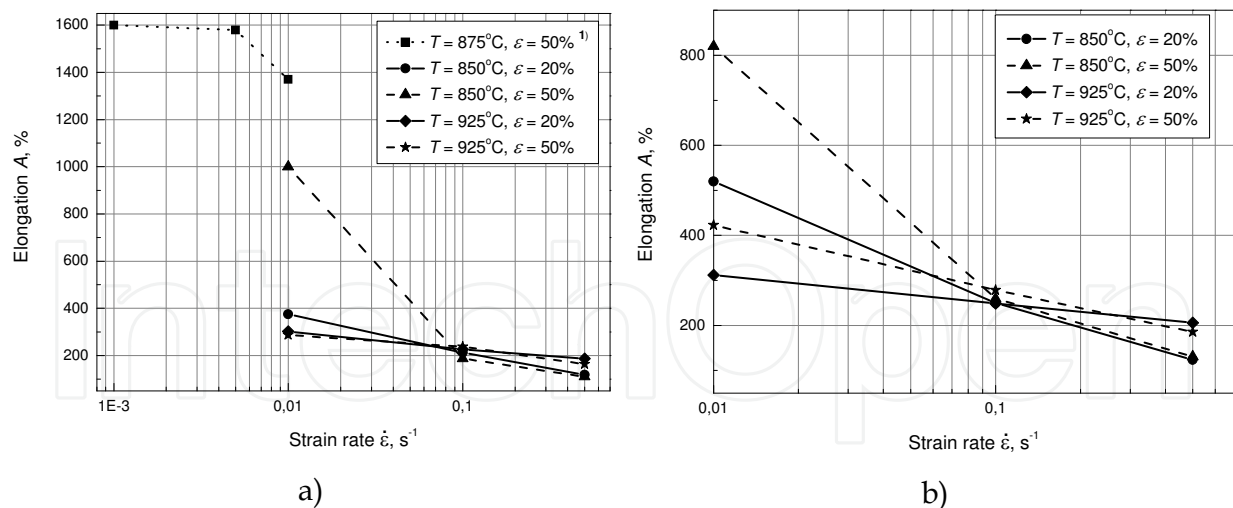
On the basis of tensile tests at 850°C and 925°C on thermomechanically processed Ti-6Al-4V and Ti-6Al-2Mo-2Cr alloys it was found that the maximum flow stress σ_{pm} decreased with growing temperature of deformation but increased with strain rate (Fig. 37). It was found that the maximum flow stress σ_{pm} determined in tensile test is higher at lower test temperature 850°C for the strain rate range applied (Fig. 37). There is no significant effect of degree of initial deformation (forging) of two investigated alloys on σ_{pm} value for both 850°C and 925°C test temperature (Fig. 37).

The relative elongation A of hot deformed Ti-6Al-4V and Ti-6Al-2Mo-2Cr titanium alloys decreased with the increasing strain rate $\dot{\varepsilon}$ in the whole range applied (Fig. 38). For strain rate $\dot{\varepsilon}$ above 0.1 the influence of forging reduction ε in thermomechanical processing and tensile test temperature is very slight. Considerable differences are visible for $\dot{\varepsilon} = 1\cdot 10^{-2} \text{ s}^{-1}$ where the maximum A value was achieved for both alloys deformed at 850°C. After thermomechanical processing TMP-II ($\varepsilon \approx 50\%$) alloys exhibit maximum elongations, typical for superplastic deformation (Fig. 38). It seems that higher grain refinement obtained in thermomechanical process enhanced hot plasticity of two-phase titanium alloys deformed with low strain rates. Similar behaviour was observed in previous works on superplasticity of thermomechanically processed Ti-6Al-4V alloy (Motyka, 2007; Motyka & Sieniawski, 2004). It was found that fragmentation and globularization of elongated α phase grains during initial stage of hot deformation restricted grain growth and resulted in higher values of total elongation in tensile test.



¹⁾ Results obtained in tensile tests in fine-grained superplasticity region for Ti-6Al-4V alloy after TMP-II processing (Motyka & Sieniawski, 2004)

Fig. 37. The σ_{pm} - $\dot{\varepsilon}$ dependence (on the basis of tensile test) for Ti-6Al-4V (a) and Ti-6Al-2Mo-2Cr (b) alloys after processing TMP-I and TMP-II (Motyka & Sieniawski, 2010)



1) Results obtained in tensile tests in fine-grained superplasticity region for Ti-6Al-4V alloy after TMP-II processing (Motyka & Sieniawski, 2004)

Fig. 38. The $A - \dot{\epsilon}$ dependence (on the basis of tensile test) for Ti-6Al-4V (a) and Ti-6Al-2Mo-2Cr (b) alloys after processing TMP-I and TMP-II (Motyka & Sieniawski, 2010)

4. Superplasticity of titanium alloys

Superplasticity is the ability of polycrystalline materials to exhibit very high value of strain (tensile elongation can be even more than 2000%), appearing in high homologous temperature under exceptionally low stress which is strongly dependent on strain rate. Generally two types of superplasticity are distinguished: fine-structure superplasticity (FSS) – considered as an internal structural feature of material and internal-stress superplasticity (ISS) caused by special external conditions (e.g. thermal or pressure cycling) generating internal structural transformations that produce high internal stresses independent on external stresses.

FSS phenomenon is observed in isotropic fine-grained metallic materials under special conditions: limited range of low strain rates and temperature above $0.4 T_m$. Main features of superplastic deformation are: high value of strain rate sensitivity parameter ($m > 0.3$), lack of strain hardening, equiaxial shape of grains not undergoing changes, conversion of texture during deformation, low activity of lattice dislocations in grains and occurrence of intensive grain boundary sliding (GBS) with associated accommodation mechanisms (Grabski, 1973; Nieh et al., 1997).

One of the titanium alloys which has been extensively studied in aspect of superplasticity is widely used Ti-6Al-4V alloy. Results concerning research on this alloy published in world scientific literature indicate meaningful progress in evaluation and applications of superplasticity in last 30 years. In the beginning of 70's maximum superplastic tensile elongation of Ti-6Al-4V alloy was about 1000% at the strain rate of 10^{-4} s⁻¹ (Grabski, 1973), whereas in few last years special thermomechanical methods were developed that enabled doubling the tensile elongation and increasing strain rate by the factor of 100 (Inagaki, 1996) (Table 4).

Alloy	Phase composition	Elongation ε [%]	Grain size \bar{d} [μm]	Temperature T [$^{\circ}\text{C}$]	Strain rate $\dot{\varepsilon}$ [s^{-1}]
Two-phase $\alpha + \beta$ alloys					
Ti-4Al-4Mo-2Sn-0.5Si (IMI550)	$\alpha + \beta$	2000	4	885	$5 \cdot 10^{-4}$
Ti-4.5Al-3V-2Mo-2Fe (SP-700)	$\alpha + \beta$	2500	2-3	750	10^{-3}
Ti-5Al-2Sn-4Zr-4Mo-2Cr-1Fe (β -CEZ)	$\alpha + \beta$	1100	2-3	72	$2 \cdot 10^{-4}$
Ti-6Al-4V	$\alpha + \beta$	2100	2	850	10^{-2}
Ti-6Al-2Sn-4Zr-2Mo	$\alpha + \beta$	2700	1-2	900	10^{-2}
Ti-6Al-2Sn-4Zr-6Mo	$\alpha + \beta$	2200	1-2	750	10^{-2}
Ti-6Al-7Nb (IMI367)	$\alpha + \beta$	300	6	900	$3 \cdot 10^{-4}$
Ti-6.5Al-3.7Mo-1.5Zr	$\alpha + \beta$	640	6-7	600	10^{-4}
Ti-6Al-2Sn-2Zr-2Mo-2Cr-0,15Si	$\alpha + \beta$	2000	4	885	$5 \cdot 10^{-4}$
Intermetallics based alloys					
Ti-24Al-11Nb	α_2 (Ti ₃ Al) + β	1280	4	970	10^{-3}
Ti-46Al-1Cr-0.2Si	γ (TiAl) + α_2 (Ti ₃ Al)	380	2-5	1050	10^{-3}
Ti-48Al-2Nb-2Cr	γ (TiAl) + α_2 (Ti ₃ Al)	350	0.3	800	$8.3 \cdot 10^{-4}$
Ti-50Al	γ (TiAl) + α_2 (Ti ₃ Al)	250	<5	900-1050	$2 \cdot 10^{-4}$ - $8.3 \cdot 10^{-3}$
Ti-10Co-4Al	$\alpha + \text{Ti}_2\text{Co}$	1000	0.5	700	$5 \cdot 10^{-2}$
Titanium matrix composites					
Ti-6Al-4V + 10%TiC	$\alpha + \text{TiC}$	270	5	870	$1.7 \cdot 10^{-4}$
Ti-6Al-4V + 10%TiN	$\alpha + \text{TiN}$	410	5	920	$1.7 \cdot 10^{-4}$

Table 4. Superplastic deformation conditions of selected titanium alloys and titanium matrix composites (Sieniawski & Motyka, 2007)

Relatively new group of superplastic titanium alloys are TiAl or Ti₃Al intermetallics based alloys (Tab. 2). It is well known that intermetallics based alloys have a high relative strength, and good high-temperature creep resistance. Widespread usage of those materials is limited mainly by their low plasticity precluding forming of structural components using conventional plastic working methods. In this case pursuit to obtain fine-grained microstructure enabling superplastic deformation seems to be very promising [Hofmann et al., 1995; Imayev et al., 1999; Kobayashi et al., 1994; Nieh et al., 1997].

Main criterion for superplastic materials is possibility of obtaining fine-grained and equiaxial microstructure. Desired microstructure is most often obtained by conventional plastic working methods coupled with suitable heat treatment and severe plastic deformation methods (i.e. equal-channel angular pressing – ECAP). Superplastic forming (SPF) of titanium alloys is limited by relatively long time and high deformation temperature. It was established that grain refinement causes increase of strain rate and decrease of superplastic deformation temperature (Fig. 39) (Sieniawski & Motyka, 2007).

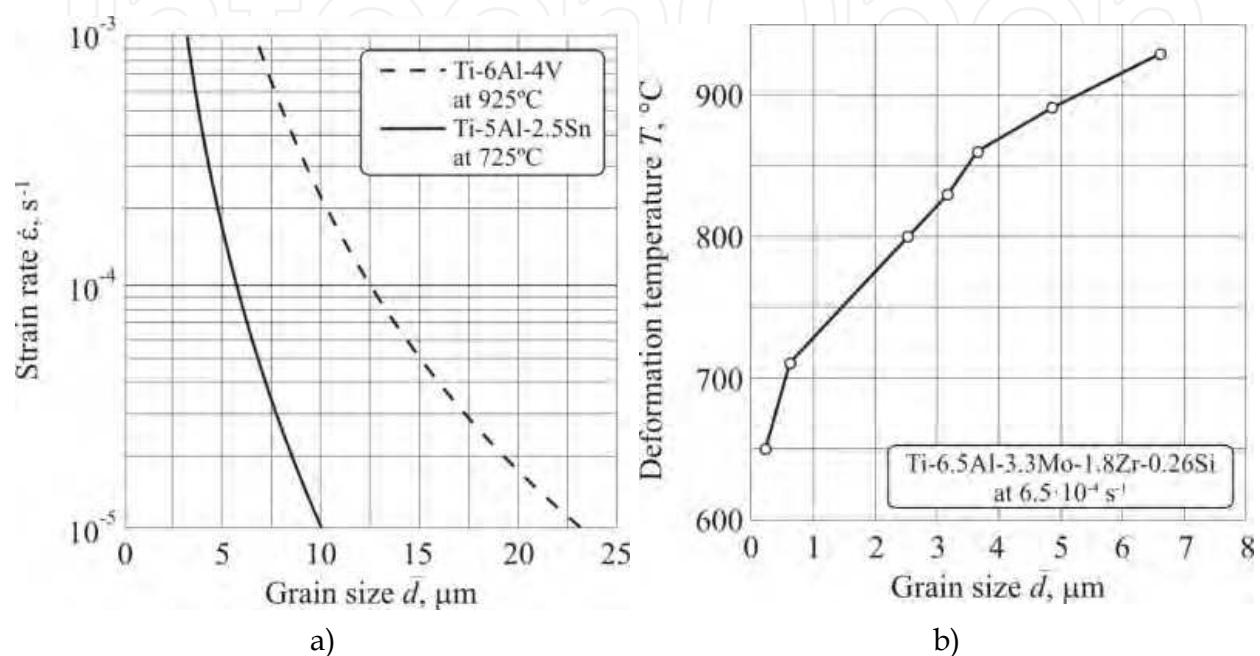


Fig. 39. Effect of grain size on strain rate of superplastic deformation for Ti-6Al-4V and Ti-5Al-2.5Sn alloys (a) and on superplastic deformation temperature for Ti-6.5Al-3.3Mo-1.8Zr-0.26Si alloy (b) (Sieniawski & Motyka, 2007)

Taking into account the mechanism of superplastic deformation equiaxed microstructure favours proceeding of GBS. It was found that in fine grained polycrystalline materials with grains elongated crosswise deformation direction GBS is limited. The main reason is difficulty of deformation accommodation in triple points. Transverse deformation is also related to cavities formation along grain boundaries and precludes superplastic deformation (Nieh et al., 1997). It is emphasised that superplastic deformation does not cause shape changes of equiaxed grains. However, gradual transformation of texture is observed what indicates that GBS plays a crucial role in superplastic deformation (Grabski, 1973; Zelin, 1996).

On the basis of the results of research works conducted at the Department of Materials Science of Rzeszow University of Technology it was found that initial microstructure of superplastic titanium alloy can be different from equiaxed one. High superplasticity was observed in Ti-6Al-4V alloy with microstructure composed of strongly elongated and deformed α grains (Fig. 40a) (Motyka & Sieniawski, 2004). It was established that during heating and first stage of superplastic deformation significant changes of the morphology of phases occur (Fig. 40b) (Motyka, 2007).

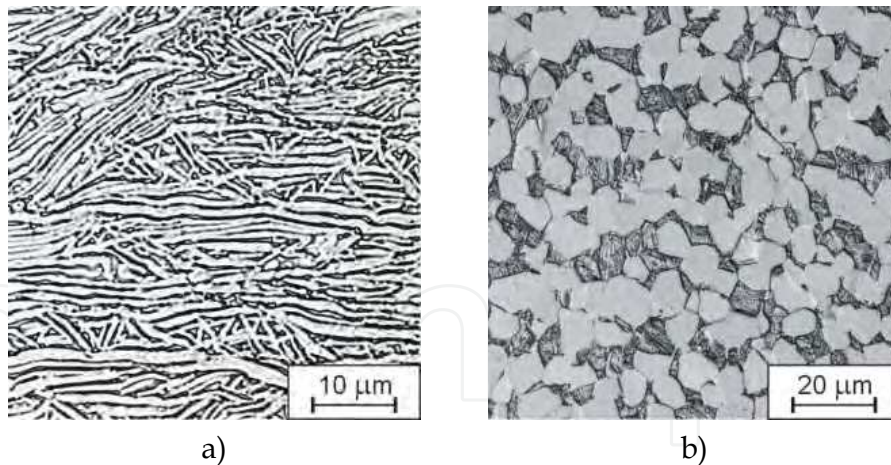


Fig. 40. Microstructure of Ti-6Al-4V alloy before (a) and after superplastic deformation (b) - temperature 850°C and strain rate of 10^{-3} s^{-1} (Sieniawski & Motyka, 2007)

Along with grain size and shape, volume fraction of particular phases in the alloy also affects its superplasticity. Properties of phases in two-phase $\alpha+\beta$ titanium alloys differ considerably. α phase (hcp) has less slip systems and two order of magnitude lower self-diffusion coefficient than β phase (bcc). These features suggest that in the superplasticity conditions α phase has a higher plasticity than β phase. It was confirmed by results obtained from experiments on superplasticity in Ti-6Al-4V alloy where deformation in α grains was observed. Density of dislocations was found to be very low in β grains (Bylica & Sieniawski, 1985; Inagaki, 1996; Jain et al., 1991, Meier et al., 1991; Nieh et al., 1997).

It was established that increase in volume fraction of β phase in alloy causes decrease of the effect of α grain size (Meier et al., 1991). Maximum values of elongation and strain rate sensitivity factor m as a function of β volume fraction is shown in Figure 41. Increase in relative volume of β phase causes improvement of superplasticity of titanium alloys. The best superplastic properties of two-phase $\alpha+\beta$ titanium alloys are achieved for 40-50% volume fraction of β phase (Nieh et al., 1997). Whereas similar properties of intermetallics based alloys are possible for about (20-30)% volume fraction of β phase (Kim et al., 1999; Lee et al., 1995).

Superplasticity of titanium alloys depends on relationship between grain growth control and plasticity. β grains are characterized by high diffusivity therefore they grow extremely rapidly at the superplastic deformation temperature which does not favour superplastic flow (Meier et al., 1992). Particular volume fraction of α phase considerably limits β grains growth because in this case long distance diffusion of alloying elements is necessary (e.g. vanadium in β phase). The second phase, besides stabilization of microstructure, influences the rate of grain boundary (α/α , β/β) and phase boundary (α/β) sliding (Inagaki, 1996; Jain et al., 1991, Meier et al., 1991; Nieh et al., 1997). Increase of volume fraction of β phase causes decrease of α/α grain boundary areas and consequently their contribution to deformation by GBS. It is thought that improvement of superplasticity of $\alpha+\beta$ titanium alloys caused by increase of volume of β phase should be considered in following aspects (Inagaki, 1996): α/β phase boundary sliding, β/β GBS and contribution of other deformation mechanisms.

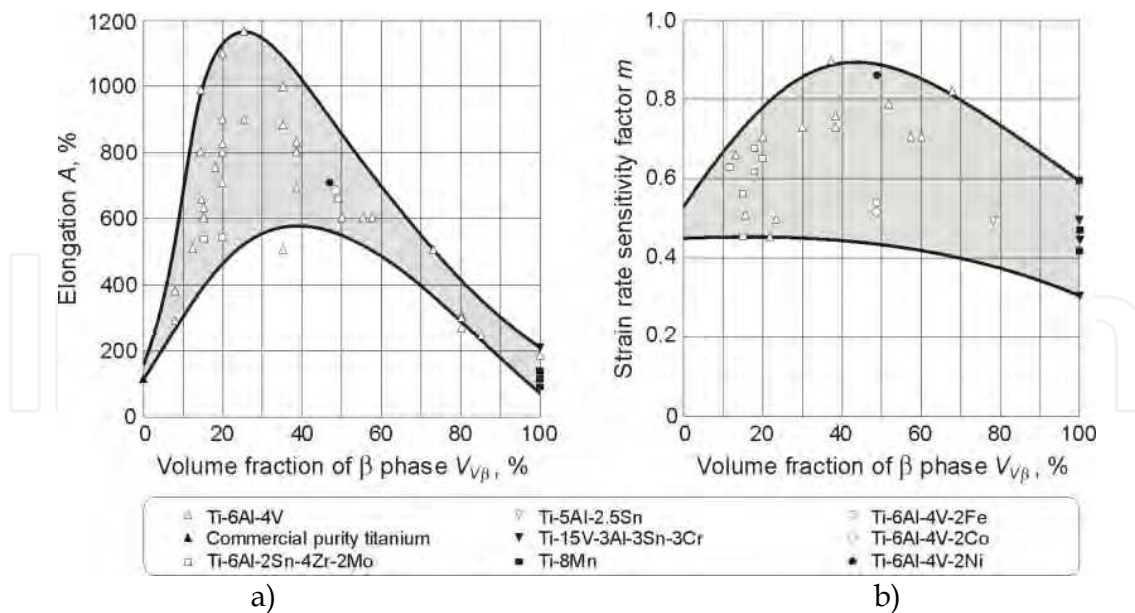


Fig. 41. Effect of volume fraction of β phase on elongation ε (a) and strain rate sensitivity m (b) in selected titanium alloys (Nieh et al., 1997)

Most often microstructure of $\alpha+\beta$ superplastic titanium alloys is composed of α and β grains which have similar size and shape. Interesting results were obtained for Ti-6Al-4V alloy where α grains were separated by thin films of β phase. Superplastic elongation in this case was more than 2000%. Further investigations indicated that during superplastic deformation thin films of β phase coagulated in triple points into larger particles having irregular forms. Thanks to that α/α grain boundaries free of β thin films were formed. It can be expected that sliding along these grain boundaries proceeds easily. However it was revealed that at this stage superplastic deformation is almost completed and deformation within grains becomes dominant deformation mechanism. It seems that α/α grain boundary sliding is not dominant superplastic deformation mechanism. In this case the effect of β phase thin film can be comparable to role of grain boundaries in single phase materials. Slip and shearing in β phase thin film is caused by movement and rotation of neighbouring α grains. Mentioned processes enable accommodation of grain boundary and phase boundary sliding (Inagaki, 1996). Other investigations also indicate accommodative role of β phase, in which substantially higher dislocations density is observed than in α phase grains. It was noticed simultaneously that dislocations density in α phase increases together with decrease in temperature and increase in strain rate of superplastic deformation (Kim et al., 1999; Meier et al., 1991). Superplasticity of titanium alloys with intermetallic phases like Ti-12Co-5Al and Ti-6Co-6Ni-5Al is observed for grain size about $0.5 \mu\text{m}$. Particles of Ti_2Co and Ti_2Ni phases (about 27% of volume) advantageously influence the grain refinement and limit grain growth during superplastic deformation (Nieh et al., 1997).

5. Conclusion

Hot plasticity of two-phase titanium alloys strongly depends on values of stereological parameters of microstructure. It is possible to develop appropriate microstructure of these alloys yielding optimum plastic flow stress and critical strain values based on

technological plasticity criterion. The value of critical strain depends on microstructure morphology and deformation conditions. The character of dependence of plastic flow stress as well as results of microstructure examination support conclusion that dynamic processes of microstructure recovery take place above the temperature range of $\alpha+\beta\leftrightarrow\beta$ phase transformation.

Thermomechanical processing enables microstructure and hot plasticity development of two-phase $\alpha+\beta$ titanium alloys. Increase in degree of initial deformation (forging) in proposed thermomechanical processing leads to formation of more elongated and refined α grains in tested $\alpha+\beta$ alloys. The most significant effect of degree of initial deformation occurs for the lowest strain rate and lower tensile test temperature used, resulting in considerable rise of elongation A .

High superplasticity of the Ti-6Al-4V alloy does not necessarily require equiaxial microstructure. Changes of the morphology of phases during heating and first stage of superplastic deformation enables superplastic behaviour of the alloy with initial microstructure composed of strongly elongated and deformed α grains.

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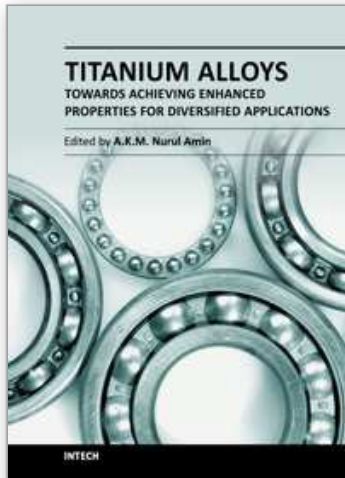
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