Influence of the parameters of the heat treatment mode on the structure formation and properties of the welded material for the alloy Ti-Al-V

E A Krivonosova, T V Olshanskaya, S N Akulova, A V Myshkina, E S Salomatova

Perm National Research Polytechnic University, 29 Komsomolsky Ave., 614990, Perm, Russian Federation

E-mail: weld-katy@mail.ru

Abstract. The article is devoted to a topical problem related to the study of the main regularities of the formation of the structure and properties of the titanium alloy Ti-Al-V of the Ti-Al-V system in additive hybrid production technologies using plasma surfacing. The aim of this work is to study the effect of various types of heat treatment on the physical and mechanical properties of the titanium alloy Ti-Al-V, obtained in the process of 3D printing.

1. Introduction

The principle of forming products by layer-by-layer growth with the use of surfacing or sintering of the material makes it possible to obtain complex-shaped parts from tools, structural steels, and other alloys. The production of such products by conventional casting methods and mechanical processing is complicated, laborious and it leads to high consumption of an expensive material, which is especially important in the production of pieces from titanium and nickel alloys [1; 19–22].

In many cases, to improve the quality of final products it is advisable to combine additive processes with traditional operations (for example, hammering, heat treatment, or machining); such technologies are usually called hybrid technologies. The simultaneous impact of various types of technological operations enhances the effect of each other and allows one to get a qualitatively new set of material properties.

Titanium alloys belong to the main structural materials being used in various industries. Their widespread application is associated with the special properties of titanium and its alloys including high specific strength, corrosion resistance, good heat resistance at operating temperatures up to 500-600°C. Titanium is well pressurized in a hot state and satisfactory in a cold state. It can be easily rolled, hammered, and stamped. Titanium and its alloys can be well welded, providing high strength and ductility of the welded joint. However, a significant disadvantage of titanium alloys in the weld and deposition technologies are some structural features, in particular, unfavourable dendritic structure, anisotropy, and banding due to transcrysalline adjustment of dendritic clusters, and, as a consequence, a decrease in mechanical properties [2], as noted in our article [22] for Cold Metal Transfer Additive Deposition.
The analysis of Russian and foreign research works shows that an increase in the physical and mechanical properties of the titanium alloys can be achieved by thermal and thermomechanical treatment, as well as by combined processing methods [2–4].

One of the most effective additive manufacturing technologies – plasma deposition – was used in this work. The use of plasma deposition provides a number of technological and economic advantages [5]. They include high productivity in the manufacture of products by layer-by-layer synthesis, control over a wide range of heat transfer to the base and deposited material and, as a result, control of the depth and width of penetration, structure, composition, and properties of the formed material.

2. Materials and Methods

Ti-Al-V welding wire was used to obtain test samples (Figure 1), the chemical composition and mechanical characteristics of the wire in accordance with GOST 27265-87 are shown in Table 1.

Table 1. Chemical composition of Ti-Al-V surfacing wire according to GOST 27265-87.

| Chemical composition, % |
|-------------------------|
| Al | V | Si | Fe | C | N | O | H | Ti – base impurities |
| 3.5–4.5 | 2.5–3.5 | 0.1 | 0.15 | 0.05 | 0.04 | 0.12 | 0.003 | 91.36–93.70 | 0.3 |

Table 2. Parameters of the plasma deposition mode of the Ti-Al-V titanium alloy in a process chamber with a controlled atmosphere.

| Polarity | I, A | U, V | V_{melting}, m/min | V_{melting}, m/h | Q_{melting}, l/min | Q_{protective}, l/min |
|----------|------|-----|---------------------|------------------|-------------------|----------------------|
| Straight | 197  | 17  | 3                   | 18               | 3 (argon)         | 8 (helium)           |

Plasma deposition was carried out in a chamber with a protective atmosphere using an EWM Tetrix 400 Plasma welding machine according to a preselected optimal surfacing mode [6], presented in Table 2.
the required level. These disadvantages can be eliminated by choosing the optimal heat treatment modes. Ti-Al-V type alloys are used in annealed and heat-hardened states. Annealing (without phase transformations, for example, recrystallization annealing below the α+β transformation temperatures) is carried out at 600-700 °C, followed by cooling in air. However, recently it has been proven that it is advisable to increase the annealing temperature up to 850...900 °C (annealing with phase recrystallization in the region above the α+β transformation), which will lead to an increase in fracture toughness and impact toughness while maintaining high plastic properties due to the formation of a mixed structure with a large proportion of the lamellar component.

The hardening mechanism of the two-phase titanium alloy Ti-Al-V during heat treatment is approximately the same as that of other metal alloys. After quenching, metastable phases (α`, α``, etc.) are formed from the β-region, which decompose upon subsequent aging (T = 500 ± 50 °C, 2...4 h) with the formation of relatively stable α- and β-phases, the number, and dispersion of which determine the obtained strength. Additional hardening of both phases is provided by neutral hardeners (tin, zirconium) and a controlled amount of elements that form interstitial solid solutions (oxygen). A high strength can be provided by adjusting the heat treatment conditions for the initial Ti-Al-V surfacing material while ensuring the required plastic characteristics.

This paper describes the investigation of the influence of the following types of heat treatment on the structure and properties of the Ti-Al-V deposited material:

1. Incomplete annealing T = 600 + 50 °C, holding time – 1...1.5 h.
2. High-temperature complete annealing T = 800 °C, holding time – 1 h.
3. Strengthening heat treatment: hardening T = 900 + 50 °C, holding time – 1 h, air cooling and subsequent aging T = 500 + 50 °C, holding time – 2...4 h.

Heat treatment was carried out in the vacuum. Polished sections were made, cut in the transverse direction of wall growth, after these types of heat treatments, to study the mechanical characteristics, macro- and microstructure. Structural studies of the deposited material, fusion zones between the layers were carried out with the use of light microscopy (Neophot 32). Quantitative metallographic analysis was performed on an automated hardware-software complex for image analysis and structural modeling “Videotest-metal” based on a metallographic microscope Altami – 587. Microhardness tests were carried out on a PMT-3 device under a load of 200 g with a step of 0.15 mm.

3. Results and Discussion
When studying the structures of the deposited material obtained in the optimal mode without heat treatment (Figure 2), the growth of primary β-grains, occurring not above 2 layers, is observed (Figure 2a). The transition between layers is visible, but it is not so clearly shown.

![Figure 2](image-url)
β-grains and between deposited layers – transition zone.

The investigation of the microstructure in accordance with X-ray phase analysis (Figure 3) showed that the structure contains the α-phase and martensite α′(α″). Partial coagulation of the α-phase occurs in the transition zone between the layers.

Taking into account that the martensite α′ phase has the same hcp lattice as the α-phase with close crystal lattice parameters, α′ and α-phases are practically indistinguishable by X-ray analysis. The α′ phase differs only by large smearing of the interference maxima. The martensitic α″-phase has an orthorhombic crystal lattice and its X-ray diffraction patterns differ from the α-phase by the splitting of some interference lines [1; 16–18]. This fact was also recorded in our article [22] for Cold Metal Transfer Additive Deposition.

![Figure 3. X-ray phase analysis of the microstructure of the Ti-Al-V deposited layer obtained in the optimal mode without heat treatment.](image)

The investigation of the macro- and microstructure of the deposited specimens after incomplete annealing, which was carried out at a temperature of 600°C, holding time being 1...1.5 hours, is shown in Figure 4.

![Figure 4. Macrostructure (a) of the deposited Ti-Al-V obtained by additive](image)
plasma deposition in a controlled atmosphere after incomplete annealing and specific areas of the microstructure (intragranular structure (b), the boundary between deposited layers and between primary β-grains (c).

After incomplete annealing at a temperature of 600 °C, no significant changes in the structure of the deposited layers are observed. The boundaries of the deposited layers are visible on the macro section. The precipitation of coagulated particles of an α-phase occurs in the transition zone between the layers. The thickness of the precipitated plates of the α-phase practically does not vary along the boundaries of primary β-grains. Martensite needles α'(α'') and thin plates of the α-phase are clearly visible in the microstructure.

Figure 5 shows the macro- and microstructure of the deposited metal after complete annealing, which was carried out at a temperature of 800 °C, holding time being 1 hour. Complete annealing at a temperature of 800 °C led to the partial decomposition of martensite α'(α'') and coagulation of the nonequilibrium α-phase, formed earlier in the structure during deposition. The transition zones between the deposited layers are practically not visible on the macro section (Figure 5) of the cross-section.

The decomposition followed a redistribution of alloying elements. This process led to the separation of an α-phase and the enrichment of α'(α'')-martensite with β-stabilizers: α'(α'') → α'(α'')-enriched + α. Distinct boundaries between the deposited layers practically disappear in the microstructure: coagulation of an α-phase of their partial globulation has occurred. The thickness of an α-phase decreases at the boundary between the former primary β-grain. The microstructure shows changes in the shape of martensite needles α'(α'')-enriched, their crushing and the thickness increasing are observed.

Investigations of the macro- and microstructure of the deposited layer after complete hardening heat treatment, which consisted of quenching from a temperature of 900 ± 50 °C, holding for 1h and subsequent aging at a temperature of 500 ± 50 °C for 2...4h, are shown in Figure 5.
Hardening with aging did not lead to the refinement of the primary $\beta$-grains, i.e. the morphology of the macrostructure of the deposited material after such heat treatment remains unchanged. Most likely, it can be explained by the fact that the phase transition of the initial phase to martensite $\alpha'(\alpha'')$ during quenching and, further, to phases ($\alpha + \beta$) during subsequent aging occurred inside the primary $\beta$-grains.

When quenching from 900 ... 950°C, the maximum amount of metastable $\beta$-phase is recorded [7, 8, 13–15]. The transitions between the deposited layers are not visible on macro section.

Heat treatment – two-stage annealing – was carried out according to the following mode: heating temperature - 850°C, holding time – 2 hours, cooling to a temperature of 600-50°C, holding time for 30 min, cooling along with a furnace, and the macro- and microstructure of the deposited metal are shown in Figure 6.

The transitions between the deposited layers are invisible in the macrostructure of the samples (Figure 6 (a)). In the microstructure (Figure 6 (b–c)), distinct boundaries between the deposited layers almost disappeared, since coagulation of the $\alpha$-phase of their partial globulation occurred. The thickness of an $\alpha$-phase decreases at the boundary between the former primary $\beta$-grain. The intracranial structure resembles a basket-like structure, the length of the $\alpha$-phase plates decreases, and their width increases. The plates have taken the so-called "branched" form, and have been divided into a chain of separate $\alpha$-particles.

At the first stage, during the heating up to 800°C, the decomposition of $\alpha'(\alpha'')$-martensite occurs according to the following scheme: 1) redistribution of alloying elements; 2) separation of the $\alpha$-phase, as a result of which $\alpha'(\alpha'')$-martensite is enriched with $\beta$-stabilizers: $\alpha'(\alpha'') \rightarrow \alpha'(\alpha'')_{\text{enriched}} + \alpha$; 3) the process ends with the partial separation of a small amount of the $\beta$-phase from $\alpha'(\alpha'')_{\text{enriched}}$ martensite.

**Figure 6.** Macrostructure of Ti-Al-V surfacing, obtained by additive plasma deposition in a controlled atmosphere with an optimized thermal cycle, after two-stage annealing, specific areas of the microstructure (boundary between primary $\beta$-grains (b), structure inside the layers (c).
At the second stage of annealing at 600°C, the process of ordering occurs in the martensitic phase $\alpha'(\alpha')_{\text{enriched}}$, leading to the formation of coherent ellipsoidal particles of $\alpha_2$-phase with a completely ordered structure, separated by an inhomogeneous $\alpha$-phase: $\alpha'(\alpha')_{\text{enriched}} \rightarrow \alpha + \alpha_2$.

Phase $\alpha_2$ is based on the Ti$_3$Al compound. This hardening is also called dispersion hardening of the $\alpha$-phase [1; 2; 7–10, 22]. The final structure $\alpha + \alpha_2 + \beta$ is confirmed by X-ray phase analysis (Figure 7).

![Figure 7. X-ray phase analysis of the Ti-Al-V deposited layer performed in a two-stage complete annealing mode.](image)

The calculated microhardness values were obtained during the statistical analysis of the microhardness data of the deposited samples before and after heat treatment (Table 3). Histograms presented in Figure 8 were constructed on the basis of these results.

|                  | Deposition on optimized mode without thermal treatment | Deposition after incomplete annealing | Deposition after high-temperature complete annealing | Deposition after hardening thermal treatment | Deposition after two-stage annealing |
|------------------|-------------------------------------------------------|--------------------------------------|-----------------------------------------------------|--------------------------------------------|------------------------------------|
| **HV$_{0.2}$, kg/mm$^2$** |                                                       |                                      |                                                     |                                            |                                    |
| Average value    | 341.5                                                 | 355                                  | 336.9                                               | 333                                        | 371.6                              |
| Minimum          | 302                                                   | 327                                  | 294                                                 | 300                                        | 324                                |
| Maximum          | 393                                                   | 390                                  | 377                                                 | 407                                        | 462                                |
| Standard deviation | 16.8                                                 | 14.9                                 | 11.3                                                | 21.2                                       | 18.7                               |

Statistical analysis shows that the average value of microhardness is minimal for the deposited samples after high-temperature complete annealing and hardening heat treatment; the standard deviation of microhardness decreases for the samples after high-temperature annealing.

The histograms represented in Figure 8 show the nature of the distribution of microhardness values, which makes the performed statistical calculations more visual. It can be seen that the samples after two-stage heat-treatment have a higher mechanical homogeneity.

Further, the deposited samples after various heat treatments were subjected to mechanical tests, the obtained data are presented in Table 4.
When comparing the mechanical properties of the deposited samples without heat treatment (Table 1) and with heat treatment, it can be concluded that annealing provides a combination of high strength, heat resistance, and good plasticity [11; 12]. In the case of incomplete annealing of the samples, the study of their mechanical properties shows that this type of heat treatment does not provide an increase in the characteristics of the deposited metal, and, apparently, leads to a decrease in residual stresses. The differences in microhardness are observed.

![Figure 8](image)

**Figure 8.** Statistical analysis of microhardness obtained by additive plasma deposition in a controlled atmosphere without heat treatment (a), with incomplete annealing (b), after high-temperature complete annealing (c), after hardening thermal treatment (d), and after two-stage annealing (e).

| Thermal treatment mode                                      | \( \sigma_a \) MPa | \( \sigma_{0.2} \) MPa | \( \delta \) % | \( \psi \) % | KCU, kgf·m/cm² |
|-------------------------------------------------------------|---------------------|------------------------|----------------|--------------|----------------|
| Incomplete annealing: T = 600 + 50°C, time – 1…1.5 h        | 870 ± 20            | 800 ± 15               | 10 ± 1         | 27 ± 10      | 6.2 ± 1.0      |
| High-temperature annealing: T = 800°C, time – 1 h           | 840 ± 20            | 770 ± 15               | 8.2 ± 1.4      | 28 ± 6       | 6.9 ± 0.7      |
| Hardening: T = 900 + 50°C, time – 1 h, air cooling + aging: T = 500 + 50°C, 2…4 h | 820 ± 14            | 740 ± 25               | 10 ± 3         | 31 ± 10      | 8 ± 1          |
| Two-stage complete annealing: I – stage T = 850°C, time – 2 h; II – stage T = 600 - 50°C, time – 30 min | 900 ± 20            | 800 ± 20               | 10.5 ± 1.2     | 33 ± 5       | 5.0 ± 0.5      |

**Table 4.** The results of the mechanical tests of the samples.
The investigations of the mechanical properties of the deposited metal after complete annealing have shown that this heat treatment mode does not provide the required characteristics. The microhardness of the samples is lower than the values obtained after incomplete annealing.

The mechanical properties of the deposited metal after hardening with aging do not provide the required values. The metal demonstrates very good impact toughness, satisfactory plastic characteristics, but insufficient strength. The microhardness values also vary.

High strength and high-temperature strength in combination with good ductility of the alloy can be achieved with the use of isothermal annealing.

The mechanical properties of the deposited metal after two-stage annealing provide the required level of mechanical characteristics. High values of plastic and strength characteristics were obtained in combination with good impact toughness.

4. Conclusion
Coagulated particles of an \(\alpha\)-phase are precipitated in the transition zone between the layers of the deposited Ti-Al-V titanium alloy after incomplete annealing at a temperature of 600°C. The microstructure contains \(\alpha'(\alpha'')\) martensite needles and thin plates of the \(\alpha\)-phase. This type of heat treatment leads to the residual stresses decrease. The strength characteristics do not improve with relation to the initial values for the deposited metal (\(\sigma_0 = 855...870\) MPa); the plasticity and impact toughness remain at a sufficiently high level (\(\delta = 9.1...10.4\%\); \(\psi = 22.8...35.4\%\); \(KCU = 5.6...6.9\) kgf \(\cdot\) m/cm\(^2\)).

Complete annealing at a temperature of 800°C of the deposited titanium alloy Ti-Al-V led to the partial decomposition of martensite \(\alpha'(\alpha'')\) and coagulation of the nonequilibrium \(\alpha\)-phase formed earlier in the structure during deposition. Such structural features of the Ti-Al-V alloy after complete annealing did not improve the mechanical characteristics with relation to the initial values of the deposited metal (ultimate strength is \(\sigma_0 = 830...850\) MPa with low plasticity).

Hardening followed by aging does not lead to crushing of the primary \(\beta\)-grains. During the aging of the hardened sample, the decomposition of martensite and metastable \(\beta\)-phase occurs with the formation of an unfavorable coarse-lamellar structure \((\alpha + \beta)\). The \(\beta\)-phase remains in an amount less than 10\% after aging of the Ti-Al-V type alloys. Such features of the structure of the Ti-Al-V alloy after hardening and aging do not provide the required mechanical properties; the heat-treated alloy has sufficiently high plastic properties \(\delta = 9.1...12.3\%\); \(\psi = 23.0...38.1\%\) and the highest level of impact toughness \(KCU = 7.3...8.5\) kgf \(\cdot\) m/cm\(^2\). At the same time, the level of ultimate strength \(\sigma_0 = 820...830\) MPa turns out to be lower than the initial values of the deposited metal and the level of technical task requirements.

Two-stage annealing with isothermal stop (holding at a temperature of 850°C, slow cooling together with a furnace to 600–50°C, holding at this temperature and subsequent slow cooling) leads to the fact that distinct boundaries between the deposited layers disappear and the structure stabilizes. This type of heat treatment provides an impact strength \(KCU = 5.0 \pm 0.5\) kgf \(\cdot\) m/cm\(^2\) and plasticity \(\delta = 10.5 \pm 1.2\%\); \(\psi = 33 \pm 5\%\) with high strength \(\sigma_0 = 900 \pm 20\) MPa.

The optimal mode of heat treatment of the VT6sv alloy obtained by 3D-printing by plasma deposition of wire material in a technological chamber with a controlled atmosphere of inert gas has been determined, being two-stage annealing, which provides the physical and mechanical properties of the obtained metal required by the Technical Specification.

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