Diffusion Parameters in Liquid Phase Sintering of the Al-Cu System

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Abstract: Within the framework of the conception of the priority of the diffusion of atoms from the liquid phase into the solid at their interface, a method of quantitative determination of diffusion parameters of interaction of solid and liquid phases by experimental curves of growth and densification of powder bodies during liquid phase sintering has been suggested. Results of the determination of diffusion parameters and volume changes of powder bodies of the Al-Cu system during sintering were produced on the basis of solution of the reverse and direct problems.

Keywords: Liquid phase sintering; Volume changes; Dilatometric curves; Diffusion parameters; Numerical computation.

Резюме: В рамках концепции о первичности процесса диффузии атомов из жидкой фазы в твердую на межфазной границе предложен метод количественного определения диффузионных параметров взаимодействия твердой и жидкой фаз по экспериментальным кривым роста и усадки порошковых тел при жидкозаполнном спекании. Представлены результаты определения диффузионных параметров и объемных изменений порошковых тел системы Al-Cu при спекании на основе решений обратной и прямой задач.

Ключевые слова: Жидкозаполнное спекание; объемные изменения; дилатометрические кривые; диффузионные параметры; численный расчет.

Содержание: У оквиру концепции приоритета диффузии атома из течной фазы в чистоте на ее границе предложен метод количественного определения диффузионных параметров реакции чистой и течной фаз на экспериментальных кривых роста и усадки твердых частиц тела током течнойфазного спекания. Результаты определения диффузионных параметров и объемных изменений твердых частиц тела системы Al-Cu током спекания показаны на основе решения обратной и прямой задач.

Ключевые речи: Течнозаполнное спекание; объемные изменения; дилатометрические кривые; диффузионные параметры; численный расчет.

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Introduction

A mathematical model of diffusion interactions of a solid phase with liquid was considered in [1-2] on the basis of the concept that the solid phase passes into the melt only after supersaturating of its surface layer with a component forming the liquid phase [3]. The model was constructed under the following assumptions:
- The interaction of components occurs on an infinite flat boundary sharing solid and liquid phases;
- The components are distributed evenly throughout the volume of the liquid phase;
- The transition of a surface layer of a solid phase of some thickness into liquid state takes place only after supersaturation of this layer with the second component up to the concentration corresponding in the phase diagram to the liquidus line at a given temperature.

Previously the author of the experimental work [4] used such an approach to qualitatively explain and quantitatively describe the dependence of volume changes of powder metal bodies of binary systems during liquid phase sintering on a number of parameters. The aim of the present work is to solve a reverse problem on the basis of analytical dependences obtained in [1-2] and experimental data [4-5], that is to determine kinetic coefficients of the diffusion interaction of the solid phase with liquid during sintering of Al–4 wt.% Cu (1.74 at.% ) compacts.

Model and analytical equations

The behaviour of a powder body during sintering can be described by the behaviour of one elementary cell reflecting its structure in this or that measure. An elementary cell of a powder compact from two components forming an eutectic system can be represented as two flat layers consisting of a solid component A and melt of an eutectic composition A+B at temperatures above the eutectic melting point (Fig. 1). In a rough approximation this elementary cell is adequate to the mathematical model [1-2].

![Fig. 1 Schematic representation of a powder body of the Al-Cu eutectic system of two elementary cells at temperatures, which are higher than the melting point of the eutectic composition.](image)

The initial thickness $x_0$ of the flat layer of the solid phase in such elementary cell is equal to half of the diameter averaged out among sizes of the particles of the component A. Both the width of the liquid layer and, consequently, the size of the whole elementary cell $X_0$ along coordinate axis $x$ depend on the total concentration $C_0$ of the component $B$ in the powder body as well as on its concentration $C_0$ in the melt. Because the quantity of the component $B$ in the system remains constant, it is possible to write:
\[ C_0(X_0 - x_0) = C_g X_0. \]
From here, the dimension of the elementary cell \( X_0 \) is related to the initial thickness of the layer of the solid component \( x_0 \) by the direct ratio:
\[ X_0 = \frac{C_0}{C_0 - C_g} x_0. \]

At the first stage of interaction of the solid and liquid phases, the interface \( x_g \) which was at the point \( x_0 \) in the initial moment of time shifts in the direction of the melt as a result of diffusion component \( B \) from the liquid phase in the solid \( A \). In the mathematical description of the process of diffusion from liquid phase in solid, the main difficulty consists in the choice of a kinetic law in accordance with which this transition proceed, as such a law has not been experimentally investigated so far. An assumption was made in [6] that the rate of transition of atoms from liquid phase into solid is constant. The kinetic equation of the interface movement was determined in [1-2] on the basis of the following assumptions. From the first Fick’s law, the amount of component \( B \) passing across the area unit of the interface is proportional to a flux or concentration gradient at this interface and to time of diffusion \( t \). As a result, according to the first Fick’s law and [7-8] it is possible to describe the diffusion process on the interface through a number of atomic jumps from liquid phase into solid as follows:
\[ dM = D_g \frac{(C_0 - C_g)}{a} dt, \]
where \( C_g \) is the concentration of the component \( B \) in the solid phase at the interface \( x_g \), \( D_g \) is the coefficient of diffusion of atoms \( B \) from the melt into the solid phase at the interface, \( a \) is the width of a diffusion barrier at the interface commensurable with an interatomic distance in a condensed phase. As \( dM = C_0 dx_g \), the velocity of the interface movement can be given as:
\[ C_0 \frac{dx_g}{dt} = Q(C_0 - C_g), \]
(1)
where \( Q = \frac{D_g}{a} \) is the factor describing the rate of transition of atoms from the liquid phase in solid at the interface. With reference to phenomenon of liquid phase sintering, the interface movement in the direction of the melt means an increase of solid phase particles in volume and, hence, an increase of the whole powder body in dimensions [4].

In the second stage of the interaction of components, the supersaturated surface layer of the solid phase is converted into liquid state by melting. This event results in changing of the direction of the interface movement for the opposite one, as the interface is shifted to the solid phase. Displacement of the interface to the direction of the solid phase during sintering brings about a reduction of the size of solid phase particles, which approach to each other under the action of capillary forces, causing densification of the powder body.

The second Fick’s law of diffusion of component \( B \) in the solid phase has the following form:
\[ \frac{\partial C}{\partial t} = D \frac{\partial^2 C}{\partial x^2}, \]
(2)
where \( C \) is a current concentration, \( D \) is the diffusion coefficient of the component \( B \) in the solid phase; \( x \) is an axis of coordinates.
Thus, the growth and densification of a powder body during liquid phase sintering can be described on the basis of the model shown in Fig. 1, taking into account the direction and value of interface displacement. In our case, the component B forming the liquid phase in the Al–Cu system by means of the contact melting is copper which passes into the melt at temperatures above the melting point of the eutectic alloy (548°C), and the solid component A is aluminium. The composition of the liquid phase formed at the melting point of the eutectic alloy corresponds to the eutectic one (Al–17.3 at.% Cu), but at higher temperatures it can deviate from this ratio to greater contents of aluminium.

At temperatures below the melting point of the eutectic composition, when the liquid phase is absent, the thickness of the component A layer in the elementary cell represented in Fig. 1 answers an average size of aluminium particles, whereas the width of the solid copper layer B is determined by its total concentration in the compact. Formation of the melt of the eutectic composition itself does not cause any volume changes of the powder body, as far from all particles of aluminium in the compact participate in contact melting with copper. The contact melting occurs locally, only around copper particles. As a result, the integrity of the compacted framework is not disturbed, for volume changes of a powder body arise only due to diffusion interaction of the melt with the whole or the main part of the surface of the solid skeleton [4]. Therefore, we shall consider that after spreading of the liquid phase over the surface of all aluminium particles, the thickness of the layer of component A in the elementary cell and its size remain the same (in other words, we neglect the volume of aluminium which has been used for formation of the eutectic melt), and the concentration of component B in the melt corresponds to its content in the eutectic alloy (17.3 at.%).

The experimental dependence of the relative change of linear dimensions of the Al–1.74 at.% Cu compact during sintering on time and temperature is shown in Fig. 2a [5]. At the temperature of 540°C that does not exceed the eutectic point (548°C), a continuous but very weak growth of the compact caused by diffusion of copper into aluminium in solid state was observed. In the interval of temperatures from 570 to 590°C, liquid phase sintering takes place, which is characterised by the fact that it is finished by an appreciable growth of the compact without any subsequent densification. In this case, the higher the sintering temperature, the greater rate of swelling compact and the larger value of growth it demonstrates.

The growth of powder bodies without subsequent densification during liquid phase sintering is characteristic for so-called transient sintering, when all components forming the liquid phase are absorbed by the solid phase [9]. The increase of the compact growth rate with sintering temperature given in Fig. 2a is explained by an increase of the diffusion mobility of atoms owing to the temperature rise. However, the reason of the dependence of the growth value on temperature in this case is not clear, because the copper content in all compacts remains constant. It is possible to assume that in this interval the sintering temperature influences not only diffusivity of copper in aluminium, but also the wettability of aluminium with melt underneath the non-reducible oxide film covering its particles. The last circumstance can also render an essential influence on the growth rate of a compact with an increase of temperature.

The value of the compact growth decreases with rising of the sintering temperature within the range of 600-630°C. This law is explained by that if the temperature rises, the marginal concentration of copper in aluminium at which aluminium can still exist in solid state decreases. Let us remember that just the concentration of copper diffused into aluminium defines the amount of the powder compact growth during liquid phase sintering [4].
Fig. 2 Experimental dependence of the relative change of the linear dimensions of Al-Cu powder bodies on time and temperature during liquid phase sintering:

a – The influence of the copper concentration on the linear dimension changes in Al-Cu powder compacts during sintering at 580°C. The copper concentration: 1 – 1.3, 2 – 1.74, 3 – 2.19, 4 – 2.64 at.%, 5 – the compact temperature, 6 – the temperature in the dilatometer tube [4];

b – The influence of temperature on the relative change of the linear dimensions of Al-1.74 at.% Cu powder compacts during liquid phase sintering [5].

As to densification within the interval 600-630°C, its rate and value appear to be higher with the increase in sintering temperature. Intensification of densification with rising of the sintering temperature is explained by enhancing of the diffusion mobility of atoms, narrowing of the concentration region of the solid phase existence and increasing of the dissolution limit of aluminium in the liquid phase [4].

Let us assume that within the interval of sintering temperatures 570-630°C the "irregular" regime of diffusion takes place. It means that supersaturation of the surface layer of an aluminium particle with copper up to concentration \( C_L^* \) enabling this layer to transform into a liquid state by melting happens before the diffusion front inside the particle reaches its middle and, hence, before the whole volume of the particle transforms into a solid solution of copper in aluminium. The numerical meanings of \( C_L^* \) were determined with the help of the liquidus line in the phase diagram of the Al-Cu system [10].

At first, we will consider the process of sintering of the Al-Cu system at the temperature of 580°C (Fig. 2). According to the set problem, it is necessary to determine coefficients \( Q \) and \( D \) in Eqs. (1) and (2), which control the velocity of the interface shift during the diffusion interaction of solid and liquid phases. For the "irregular" mode of diffusion, the factors \( Q, D \) and the coordinate of moving interface \( x_g \) are related with each other in the stage of growth by the following equation [1]:

\[
x_g - x_0 + \frac{(2^{3/2})}{3} (x_g - x_0) \left( \sqrt[3]{\frac{Q}{D}} \right) = Q(t - t_0),
\]

where \( x_0 \) is the initial position of the interface between an average aluminium particle and the liquid phase in the elementary cell, \( t_0 \) is the time of the beginning of powder body growth in volume as a result of the diffusion interaction of the melt with aluminium.
By analogy with [11], let us express the relative linear thickening of the flat elementary cell \( \Delta l/l_0 \) through its parameters. Because of interface shifting in the direction of melt in the cell as well as the growth of the powder body occur only at the expense of transition of copper atoms in aluminium, it is possible to write:

\[
\Delta l/l_0 = \frac{(x_g - x_0)}{x_0} C_0. \tag{4}
\]

The coefficient \( Q \) can be determined by means of the angle of the kinetic curve inclination in Fig. 2 for sintering at 580°C and \( t = t_0 \), assuming that the concentration of solid phase \( C_g = 0 \) at the interface for \( t = t_0 \). Taking into consideration Eqs. (1) and (4), one can obtain:

\[
Q = \frac{x_0}{C_0} \frac{d(\Delta l/l_0)}{dt} \bigg|_{t=t_0}. \tag{5}
\]

The coefficient of diffusion \( D \) is determined from Eq. (3):

\[
D = \frac{8Qx_0^3(\Delta l/l_0)^3}{9[C_0Q(t-t_0) - x_0(\Delta l/l_0)]^2C_0}. \tag{6}
\]

Through the maximal increase of the thickness of the solid phase layer in the elementary cell \( \Delta l_m/l_0 \) and time \( t_m \) at which it is reached, substituting (5) in (6), we obtained the following expression for the coefficient of diffusion:

\[
D = \frac{8}{9} \frac{x_0^2}{C_0^2} \frac{d(\Delta l/l_0)}{dt} \bigg|_{t=t_0} \frac{(\Delta l_m/l_0)^3}{(t_m - t_0) - \Delta l_m/l_0}. \tag{7}
\]

Let us consider the second way of determination of the kinetic \( Q \) and \( D \) parameters. We assume that the point of the maximal increase of the solid phase layer thickness in the elementary cell is also the point at which the direction of the interface movement changes its sign for the inverse one. Accordingly, the maximal increase of the whole powder body is replaced by its densification owing to the beginning of transition of the solid phase into liquid. Then, in accordance with [1], using (4) it is possible to write:

\[
\Delta l_m/l_0 = \left( \frac{C'_S}{C_0 - C'_L} \right)^2 \frac{D C_0}{2Qx_0}, \tag{8}
\]

where \( C'_L \) is the concentration of component \( B \) in the solid phase at the interface, at which a transition of solid phase into liquid state by melting occurs.

Using (6) and (8), we obtain:

\[
Q = \frac{(3C_0 - C'_S)(\Delta l_m/l_0)x_0}{3(C_0 - C'_L)C_0(t_m - t_0)}, \tag{9}
\]

\[
D = \frac{2(3C_0 - C'_L)(C_0 - C'_L)(\Delta l_m/l_0)^2x_0^2}{3C_0^2C'_L^2(t_m - t_0)}. \tag{10}
\]
In Eqs. (9) and (10) kinetic coefficients are determined through the averaged size $2x_0$ of the solid component particles. However, it does not at all mean that the kinetic Q and D coefficients in general really depend on the particle size. Such dependence is caused by the choice of the diffusion regime within the solid phase. In this case it is the "irregular" diffusion regime, which is defined by an average size of solid phase particles. It follows from here that when solving a reverse problem of finding kinetic coefficients on the basis of volume changes of powder bodies during sintering, correct knowledge on the average initial size of solid phase particles $2x_0$ used in the experiment plays an important role.

At the stage of growth of powder bodies, the curves 1-4 in Fig. 2b corresponding to the copper concentration in the mixes from 3 up to 6 wt.\% were not used for determination of coefficients Q and D, as swelling of compacts in these experiments occurred in non-isothermal conditions, when the compact temperature was not reached yet 580°C. The parameters Q and D were determined, using the kinetic curves 3 and 4, in the stage of densification, which mainly proceeded at 580°C. In this case the absolute meaning of $C_L^*$ was supposed to be equal to the copper concentration in the melt of the eutectic composition, that is 17.3 at.\%.

The choice of the size value $2x_0$ for initial aluminium particles in the experiments [4-5] introduces the most serious error in estimation of the coefficients Q and D by Eqs. (8) and (9). Though in real powder systems some distribution of particles in sizes takes place, with the purpose of essential simplification of the accounts an average size of the Al particles was used as $2x_0$. Some error in results of calculations of parameters Q and D by Eqs. (8) and (9) is also brought in by an inaccuracy in the determination of time of completion for the process of compact growth, using the above experimental curves. The dependence of the maximal value of compact growth on sintering temperature in the interval of 570-590°C can specify that owing to the presence of oxide film on the aluminium particles and formation of tight contacts between them under compaction, not all particles are wetted with the melt over their complete surfaces. For complete wetting the maximum growth caused by absorption of all copper from the melt as a result of sintering in the temperature interval of 570-590°C should be identical, since the content of copper in the compacts is the same (1.74 at.\%). In this case the sintering temperature should only influence the rate and duration of swelling.

Results of computation

In Tab. 1 the averaged numerical values of diffusion coefficients in the Al-1.74 at.\% Cu mixture at the temperature of 580°C are given. The data were obtained on the basis of dilatometric curves shown in Fig. 2, which demonstrate changes of the compacts in linear dimension during liquid phase sintering. For comparison of the calculated diffusivity D with literary data, the values of the diffusion coefficient of copper in aluminium at 580°C published in [12-15] are presented in Tab. 1. The comparison shows that the coefficient of diffusion computed on the basis of experimental curves corresponds by order of magnitude to the values published in literature.

Thus, the mathematical model of diffusion interaction of solid and liquid phases [1-2] proposed on the basis of new ideas on the process [3], using changes of powder compacts in volume during liquid phase sintering, enables determination of coefficient Q which controls the rate of atomic transition from liquid phase into solid.
Tab. I Averaged-numerical values of diffusion coefficients $Q$ and $D$ calculated by kinetic curves in Fig. 2 at $x_0 = 50 \, \mu m$.

<table>
<thead>
<tr>
<th>Sintering stage</th>
<th>Diffusion parameter</th>
<th>Reference</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>$Q, \text{ cm/sec}$</td>
<td>$D, \text{ cm}^2/\text{sec}$</td>
</tr>
<tr>
<td>Growth</td>
<td>$3.1 \cdot 10^{-7}$</td>
<td>$1.1 \cdot 10^{-9}$</td>
</tr>
<tr>
<td>Densification</td>
<td>$8.5 \cdot 10^{-7}$</td>
<td>$3 \cdot 10^{-9}$</td>
</tr>
<tr>
<td>Literary data</td>
<td>$-3.5 \cdot 10^{-10}$</td>
<td>$1.4 \cdot 10^{-9}$</td>
</tr>
</tbody>
</table>

The processing of experimental curves in Fig. 2 has also allowed calculation of pre-exponential factors $Q_0$ and $D_0$, the activation energy for transition of copper atoms from the melt into aluminium $E_Q$ and activation energy for diffusion of copper in the aluminium lattice $E_D$. In connection with the inaccuracies mentioned above in determination of the growth stage duration caused by incomplete wetting of aluminium particles with the eutectic melt at low sintering temperatures, the calculation of kinetic parameters was carried out on the basis of compact changes in dimension at the densification stage. For this purpose the equation of interface movement in the elementary cell at the stage of transition of the solid component into liquid [2] was used. It was supposed that for the initial moment of time the regime of solid state dissolution of copper in aluminium is irregular, and the concentration of copper in the melt is close to the eutectic composition of the Al-Cu system. Then, it is possible to write:

$$ (C_0 - C_i^*)Q = -C_i^* \frac{dx_i}{dt} .$$  

(11)

Factor $Q$ in Eq. (11) through relative changes of the thickness of the solid phase layer can be given in the form:

$$ Q = \frac{C_i^* x_0}{C_0 (C_0 - C_i^*)} \frac{d(\Delta l/l_0)}{dt} .$$  

(12)

Then, assuming that $Q = Q_0 \exp(-E_Q / RT)$, the energy of activation $E_Q$ was calculated by the formula:

$$ E_Q = \frac{RT}{T_i - T_k} \ln \left\{ \frac{d(\Delta l/l_0)}{dt} \right|_{i=0} (C_0 - C_{i,0})C_{i,0}^* \right\} \frac{d(\Delta l/l_0)}{dt} \right|_{i=\infty} (C_0 - C_{i,\infty})C_{i,\infty}^* .$$  

(13)

where $T_i$ and $T_k$ are sintering temperatures at i, k = 600, 610, 620 and 630°C; $C_{i,0}^*$ and $C_{i,\infty}^*$ are solubility limits of aluminium in the melt at the appropriate temperatures.

The pre-exponential coefficient $Q_0$ was determined by the formula:

$$ Q_0 = Q \exp \left( \frac{E_Q}{RT} \right) .$$  

(14)

The energy of activation for diffusion of copper in aluminium $E_D$ was calculated using the time of the beginning of transition of the solid phase in the melt, which one can judge by the point of replacement of growth by densification. Using (8) and assuming that $D = D_0 \exp(-E_D / RT)$, it is possible to write:
\[ E_D = E_0 + \frac{T_k}{T_k - T_c} \frac{\Delta l_m}{l_0} c_{\text{Li}}^2 (C_0 - C_{\text{i,k}}^*)^2 \]

(15)

The pre-exponential coefficient \( D_0 \) was calculated using equation:

\[ D_0 = D_i \exp \left( \frac{E_D}{RT_c} \right). \]

(16)

Values of activation energy and pre-exponential factor calculated according to Eqs. (13)–(16) were averaged. The obtained results are presented in Tabs. II and III. As the differences in sintering temperatures for such accounts have appeared to be insignificant, the energy of activation and, consequently, pre-exponential coefficients are determined with a large error.

**Tab. II** Average values of pre-exponential coefficients \( Q_0 \) and \( D_0 \) at \( x_0 = 50 \, \mu m \) calculated on the basis of experimental curves in Fig. 2.

<table>
<thead>
<tr>
<th>Sintering stage</th>
<th>Diffusion parameter</th>
<th>Reference</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>( Q_0 ), cm/sec</td>
<td>( D_0 ), cm³/sec</td>
</tr>
<tr>
<td>Growth</td>
<td>3.5 \times 10³</td>
<td>20.0 \times 10³</td>
</tr>
<tr>
<td>Densification</td>
<td>8.7 \times 10⁴</td>
<td>5.5 \times 10⁴</td>
</tr>
<tr>
<td>Literary data</td>
<td>–</td>
<td>0.6-2.3</td>
</tr>
</tbody>
</table>

**Tab. III** Average values of activation energy \( E_0 \) and \( E_D \) at \( x_0 = 50 \, \mu m \) calculated on the basis of experimental curves in Fig. 2.

<table>
<thead>
<tr>
<th>Sintering stage</th>
<th>Diffusion parameter</th>
<th>Reference</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>( E_0 ), kJ/mol</td>
<td>( E_D ), kJ/mol</td>
</tr>
<tr>
<td>Densification</td>
<td>99.5 ± 75</td>
<td>201.1 ± 92</td>
</tr>
<tr>
<td>Literary data</td>
<td>–</td>
<td>136.3 – 145.9</td>
</tr>
</tbody>
</table>

Numerical computations of linear dimension changes of the powder bodies of the Al-Cu system in time have been carried out on the basis of a mathematical model [1-2], using obtained diffusion coefficients. For quantitative description of these changes at the growth stage Eqs. (1) and (2) were used for the following boundary and initial conditions [1]:

\[ t \leq t_0 : \frac{dC}{dx} = 1 \text{ at } 0 \leq x \leq x_0; \]
\[ C(x) = C_0 \text{ at } x_0 \leq x \leq X_0. \]

\[ t > t_0 : \left. \frac{dC}{dx} \right|_{x=0} = 0; \left. \frac{dC}{dx} \right|_{x=x_0} = Q(C_0 - C_0'^*); \]
\[ C(x) = C_0 \text{ at } x_0 \leq x \leq X_0. \]

(17)

Besides, it was taken into account that in the initial state the liquid phase is the eutectic melt with the concentration of copper equal to \( C_0 \).

It was supposed that at the stage of densification the concentration in liquid phase, \( C_L \), is distributed evenly and the boundary conditions have the following form [2]:

\[ \left. \frac{dC}{dx} \right|_{x=0} = 0; \quad C(x_{x_0}) = C_L'^*. \]

(18)
At the densification stage the distribution of concentration in the solid phase is determined by the equation of diffusion (2), and the rate of interface movement is described by the following equation [2]:

\[
\frac{d}{dt} \int_0^z C(x)dx + C_L^* \frac{dx}{dt} = Q(C_L - C_L^*). \tag{19}
\]

Results of quantitative computations of the dependence of the linear dimension of a powder body on sintering temperature are given in Fig. 3. One can see that at low sintering temperatures only growth of the powder mixture is observed. After the stage of powder compact growth, the densification stage arises with an increase of sintering temperature. Such behaviour of the powder body is qualitatively in agreement with experimental data shown in Fig. 2, and is explained by the fact that at increased sintering temperatures supersaturation on the interface occurs at a lower concentration of the second component. In this case during saturation, the component forming the liquid phase has no time to diffuse wholly into the solid phase, which results in transition of the solid component into liquid state.

![Fig. 3 Computed dependence of changes of the linear dimensions of the Al-1.74 at.% Cu powder compacts on time and temperature during liquid phase sintering.](image)

Conclusions

1. A method of determination of kinetic parameters of the diffusion interaction of solid and liquid phases in non-equilibrium state on the basis of dilatometric curves of changes of the linear dimensions of powder bodies of binary mixes during liquid phase sintering is suggested.

2. A quantitative estimation of the coefficient controlling the rate of transition of copper atoms from the melt of the eutectic composition into aluminium particles during sintering of the Al-1.74 at.% Cu powder mixture is carried out.

3. Conducted numerical calculations of relative changes in time of the linear dimensions of compact of Al-1.74 at.% Cu mixture during liquid phase sintering using determined in the work kinetic coefficients have shown that the duration of growth and densification stages is qualitatively in agreement with an experiment.
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