Reactive Diffusion between Ag–5Pt Alloy and Sn at Solid-State Temperatures

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The kinetics of the reactive diffusion between a binary Ag–5 at% Pt alloy and Sn was experimentally studied at solid-state temperatures using Sn/Ag0.05Pt0.05/Sn diffusion couples. The diffusion couple was prepared by a diffusion bonding technique and then isothermally annealed at temperatures of \( T = 433, 453 \) and 473 K for various times up to 408 h in an oil bath with silicone oil. During annealing, an intermetallic layer of Ag3Sn dispersed with fine particles of PtSn is formed at each interface in the diffusion couple. The mean thickness of the intermetallic layer is expressed as a power function of the annealing time. The exponent of the power function takes values around 0.35 at \( T = 433-473 \) K. Thus, the growth of the intermetallic layer is controlled by grain boundary diffusion as well as volume diffusion. The growth rate of the intermetallic layer is greater for the Sn/Ag0.05Pt0.05/Sn diffusion couple than for the Sn/Ag/Sn diffusion couple at \( T = 433-473 \) K. However, the temperature dependence of the growth rate shows that the growth rate may be smaller for the Sn/Ag0.05Pt0.05/Sn diffusion couple than for the Sn/Ag/Sn diffusion couple at \( T = 393 \) K. Thus, addition of 5 at% Pt into Ag accelerates the kinetics of the reactive diffusion between Ag and Sn at \( T > 400 \) K but decelerates the kinetics at \( T < 400 \) K.

Keywords: diffusion bonding, intermetallic compounds, bulk diffusion, solder, conductor

1. Introduction

Copper-base conductor alloys are widely used in electronic industry. In order to improve the corrosion resistance, however, the Cu-base alloy is usually plated with a Au layer. If the Cu-base alloy plated with the Au layer is interconnected with a Sn-base solder alloy, Au–Sn compounds are formed at the interconnection during soldering and then grow during usual energization heating on condition that the Au layer is sufficiently thick. The Au–Sn compounds are brittle and possess high electrical resistivities, and hence their growth deteriorates the electrical and mechanical properties of the interconnection. The reactive diffusion between the Sn-base solder and Cu-base conductor alloys was experimentally studied at various solid-state temperatures by many investigators.1–15) In most of the experiments, however, diffusion couples were prepared from the solder and conductor alloys by a soldering technique. In the soldering technique, compound layers with certain thicknesses are formed at the interface between the solder and conductor alloys during soldering. The formation of the compound layers before annealing will affect their growth behavior during annealing at solid-state temperatures. Therefore, the soldering technique may not be appropriate to observation of the kinetics for the reactive diffusion at solid-state temperatures.

The solid-state reactive diffusion was experimentally observed for various binary and ternary systems in previous studies.16–35) For the binary Au/Sn system, Sn/Au/Sn diffusion couples were prepared by a diffusion bonding technique and then isothermally annealed at various solid-state temperatures.16–19) Due to annealing, AuSn2, AuSn and AuSn compound layers are produced at the Au/Sn interface in the diffusion couple at temperatures between \( T = 393 \) and 473 K. The total thickness of the Au–Sn compound layers is mathematically expressed as a power function of the annealing time, and the exponent of the power function is equal to 0.48, 0.42, 0.39 and 0.36 at \( T = 393, 433, 453 \) and 473 K, respectively.17–19) Thus, the exponent monotonically increases from 0.36 to 0.48 with decreasing annealing temperature from 473 K to 393 K.

A similar experiment was carried out for the reactive diffusion in the binary Ag/Sn system in a previous study.20) In that experiment, Sn/Ag/Sn diffusion couples were prepared by the diffusion bonding technique and then isothermally annealed at temperatures of \( T = 433-473 \) K. Owing to annealing, compound layers consisting of Ag3Sn and the \( \gamma \) phase are formed at the Ag/Sn interface in the diffusion couple. Like the Au–Sn compound layers, the total thickness of the Ag–Sn compound layers is described as a power function of the annealing time. The exponent of the power function is equal to 0.40, 0.38 and 0.36 at \( T = 433, 453 \) and 473 K, respectively. Thus, the temperature dependence of the exponent is almost equivalent for the binary Ag/Sn and Au/Sn systems. However, the overall growth rate of the compound layers is one order of magnitude smaller for the binary Ag/Sn system than for the binary Au/Sn system.17–20)

As well as Ag and Au, binary Ag–Au alloys are used as corrosion-resistant conductor materials in electronic industry. In the binary Ag–Au system,36) the complete solid-solution phase with the face-centered cubic (fcc) structure exists in a wide temperature range. If the Ag–Au conductor alloy is interconnected with the Sn-base solder alloy and then heated at solid-state temperatures under usual energization conditions, Ag–Sn and Au–Sn compounds are produced at the interconnection and then grow gradually. Such growth also deteriorates the electrical and mechanical properties of the interconnection. In order to examine the growth behavior of the Ag–Sn and Au–Sn compounds at the interconnection, the reactive diffusion in the ternary (Ag–Au)/Sn system was experimentally observed in previous studies.22,23) Here, the notation (A–B)/C means that reactive diffusion occurs between a binary A–B alloy and a pure C metal at appropriate annealing temperatures. In those experiments, Sn/Ag1-x–Au/x/
Sn diffusion couples with $z = 0.13 - 0.75$ were isothermally annealed at temperatures of $T = 393 - 473$ K for various times. Here, $z$ is the mol fraction of Au in the Ag–Au alloy. At $z = 0.37 - 0.75$, AuSn$_4$ and AuSn$_2$ compound layers dispersed with fine particles of Ag$_5$Sn are formed at the interface in the diffusion couple during annealing. However, the overall growth rate is almost identical between the Au–Sn compound layers with the Ag$_5$Sn particle at $z = 0.37 - 0.75$ and those without the Ag$_5$Sn particle at $z = 1$. On the other hand, at $z = 0.13 - 0.25$, Ag$_3$Sn and AuSn$_4$ compound layers are formed by annealing at the interface in the diffusion couple. The thickness is close to each other between the Ag$_5$Sn and AuSn$_4$ layers at $z = 0.25$ but much smaller for the AuSn$_4$ layer than for the Ag$_5$Sn layer at $z = 0.13$. The overall growth rate of the compound layers is smaller at $z = 0.25$ than at $z = 0.37 - 1$ but greater at $z = 0.25$ than at $z = 0$. Thus, the growth rate seems to decrease monotonically with decreasing composition $z$. However, the growth rate becomes slightly smaller at $z = 0.13$ than at $z = 0$. This means that addition of Au with 13 at% into Ag decelerates the growth of compounds at the interconnection between Ag and Sn during energization heating.

Since addition of Pt as well as that of Au improves the corrosion-resistance of Ag, Ag–Pt alloys can be used as corrosion-resistant conductor materials like Ag–Au alloys. Also at the interconnection between the Ag–Pt conductor and Sn-base solder alloys, however, certain compounds are produced during energization heating. Considering the Sn-base solder alloys, however, certain compounds are produced during energization heating. 37) Considering the Sn-base solder alloys, however, certain compounds are produced during energization heating. 37) Considering the Sn-base solder alloys, however, certain compounds are produced during energization heating. 37) Considering the Sn-base solder alloys, however, certain compounds are produced during energization heating.

Polycrystalline plate specimens of pure Sn with a size of $12 \text{ mm} \times 5 \text{ mm} \times 2 \text{ mm}$ were prepared by cold rolling and spark erosion from a commercial 1 kg rectangular ingot of pure Sn with purity of 99.99%. The cold-rolled plate specimens were separately annealed in evacuated silica capsules at 473 K for 2 h, followed by air cooling without breaking the capsules. The annealed Sn plate specimens were chemically polished in an etchant consisting of 20 vol% of nitric acid, 20 vol% of hydrochloric acid and 60 vol% of distilled water. The two surfaces with an area of $12 \text{ mm} \times 5 \text{ mm}$ of each Sn plate specimen were mechanically polished on 800 emery paper. One of the two polished surfaces was again mechanically polished on 1000–4000 emery papers until a depth of 100 μm and then finished using diamond with a diameter of 1 μm.

After chemical polishing, a Ag–Pt sheet specimen was immediately sandwiched between the finished surfaces of two freshly prepared Sn plate specimens in ethanol by a technique used in previous studies. 17–19) The Sn/(Ag–Pt)/Sn couples were completely dried and then isothermally heated for diffusion bonding in an oil bath with silicone oil at temperatures of $T = 433, 453$ and 473 K for times of 20, 8 and 2 h, respectively. After heating, the diffusion couples were isothermally annealed at $T = 433, 453$ and 473 K for various times up to 400 h. The summation of the heating and annealing times is hereafter merely called the annealing time $t$. Cross-sections of the annealed diffusion couples were mechanically polished using diamond with diameters of 15, 3 and 1 μm and then finished with an OP-S liquid by Struers Ltd. The microstructure of the cross-section was observed with an optical microscope (OM). The observation of the microstructure was carried out also with a back-scattered electron image (BEI) by scanning electron microscopy (SEM). Concentrations of Ag, Pt and Sn in each phase on the cross-section were measured by electron probe micro-analysis (EPMA).

3. Results and Discussion

3.1 Microstructure

A typical micrograph of OM for the cross-section of the annealed diffusion couple is shown in Fig. 1. This figure indicates the micrograph for the diffusion couple with $T = 473$ K and $t = 8$ h (2.88 × 10$^4$ s). In Fig. 1, the regions on the upper and lower sides are Sn, and the horizontal band with a thickness of about 30 μm is the Ag–Pt alloy. As can be seen in this figure, an intermetallic layer with a thickness of about 50 μm is formed at each interface between Sn and the Ag–Pt alloy. In order to identify the intermetallic layer, concentration profiles of Ag, Pt and Sn were measured by EPMA along the direction normal to the interface. A result for the diffusion couple with $T = 473$ K and $t = 17.5$ h (6.3 × 10$^4$ s) is shown in Fig. 2. In this figure, the ordinate and the abscissa indicate the mol fraction $y_i$ of component $i$ ($i = \text{Ag, Pt, Sn}$) and the distance $x$, respectively, and open squares, rhombuses and circles show the mol fractions $y_{\text{Ag}}$, $y_{\text{Pt}}$ and $y_{\text{Sn}}$, respectively. According to a recent phase diagram in the binary Ag–Pt system, 38) the composition of the Ag–Pt alloy is located in the single-phase region of the primary
solid-solution (α) phase of Ag with the fcc structure at $T = 1223$ K. As can be seen in Fig. 2, the homogeneous single-phase microstructure of the α phase was realized for the Ag–Pt alloy by annealing at $T = 1223$ K for $t = 1440$ h ($5.18 \times 10^6$ s). On the other hand, for the intermetallic layer, the plotted points are considerably scattered, and thus the identification cannot be carried out conclusively. The result in Fig. 2 is plotted as a diffusion path with open circles in a composition triangle of the ternary Ag–Pt–Sn system. Open rhombuses show the initial compositions of Ag–5Pt alloy and Sn, and open squares show the compositions of Ag$_3$Sn and PtSn$_4$. Consequently, it is concluded that the Ag$_3$Sn layer dispersed with PtSn$_4$ particles is formed at the interface in the Sn/Ag$_{90.95}$Pt$_{0.05}$/Sn diffusion couple during annealing at $T = 433–473$ K. This yields that the
three-phase equilibrium of Sn + PtSn₄ + Ag₃Sn appears at 
T = 433–473 K in the phase diagram of the ternary Ag–Pt–
Sn system.

3.2 Growth behavior of intermetallic layer

From the OM micrographs like Fig. 1, the mean thickness \( l \) of the intermetallic layer was evaluated by the following equation at each annealing time:

\[
l = A/w,
\]

where \( w \) and \( A \) are the total length parallel to the interface and the total area of the intermetallic layer, respectively, on the cross-section. The result is shown in Fig. 5. In this figure, the ordinate indicates the logarithm of the thickness \( l \), and the abscissa shows the logarithm of the annealing time \( t \). Open rhombuses, squares and circles indicate the results at \( T = 433 \), 453 and 473 K, respectively. As can be seen, the thickness \( l \) monotonically increases with increasing annealing time \( t \). The plotted points at each temperature are located well on the corresponding straight line. This means that \( l \) is mathematically expressed as a power function of \( t \) by the equation

\[
l = k(t/t₀)^n.
\]

Here, \( t₀ \) is unit time, 1 s. It is adopted to make the ratio \( t/t₀ \) dimensionless. The proportionality coefficient \( k \) has the same dimension as the thickness \( l \), but the exponent \( n \) is dimensionless. From the plotted points in Fig. 5, \( k \) and \( n \) were determined by the least-squares method. Their values are shown in Fig. 5. Using these values of \( k \) and \( n \), \( l \) was calculated as a function of \( t \) from eq. (2). The result is indicated as solid lines in Fig. 5.

The plotted points at \( T = 433, 453 \) and 473 K in Fig. 5 are shown again as open circles with solid lines in Fig. 6(a), (b) and (c), respectively. As mentioned in Sect. 1, the kinetics of solid-state reactive diffusion was experimentally observed for the ternary (Ag–Au)/Sn system and the binary Au/Sn systems in previous studies. In those experiments, Sn/Agₐ₋₁Au/Sn diffusion couples with \( z = 0–1 \) were prepared by the diffusion bonding technique and then isothermally annealed at solid-state temperatures. Here, \( z \) is the mol fraction of Au in the Ag–Au alloy. In Fig. 6, the results of the Sn/Ag/Sn, Sn/Ag₀.₈₇Au₀.₁₃/Sn and Sn/Au/Sn diffusion couples are shown again as open squares, open rhombuses and open circles with dotted lines in Fig. 6(a), respectively. Hereafter, the Sn/Ag/Sn, Sn/Au/Sn, Sn/Ag₀.₈₇Au₀.₁₃/Sn and Sn/Ag₀.₉₅Pt₀.₀₅/Sn diffusion couples are called diffusion couples AG, AU, GU and AP, respectively. As can be seen in Fig. 6, the growth rate of the intermetallic layer is one order of magnitude greater for diffusion couple AU than for diffusion couple AG but slightly smaller for diffusion couple GU than for diffusion couple AG. Therefore, the minimum growth rate is realized at \( z = 0.13 \). Here, the intermetallic layer is composed of AuSn₄, AuSn₂ and AuSn for diffusion couple AU, but mainly of Ag₃Sn for diffusion couples AG and GU. Consequently, addition of Au with 13 at% into Ag decelerates the kinetics of the reactive diffusion between Ag and Sn at \( T = 433–473 \) K. On the other hand, at \( T = 433 \) and 453 K, the growth rate of the intermetallic layer is greater for diffusion couple AP than for diffusion couple AG but smaller for diffusion couple AP than for diffusion couple AU. At \( T = 473 \) K, however, the growth rate becomes greater for diffusion couple AP than even for diffusion couple AU. Hence, addition of Pt with 5 at% into Ag accelerates the kinetics of the reactive diffusion between Ag and Sn at \( T = 433–473 \) K. The acceleration effect of Pt becomes more remarkable with increasing annealing temperature.

3.3 Rate-controlling process

The exponent \( n \) of diffusion couple AP is plotted as open circles against the annealing temperature \( T \) in Fig. 7. In this figure, the results of diffusion couples AG, GU and AL are also represented as open squares, triangles and rhombuses, respectively. If the reactive diffusion is controlled by the volume diffusion of the constituent components in each phase, the exponent \( n \) is equal to 0.5. In contrast, the grain boundary diffusion across the intermetallic layer will...
govern the reactive diffusion at low temperatures where the volume diffusion is practically frozen out. When the reactive diffusion is purely controlled by the grain boundary diffusion and grain growth occurs in the intermetallic layer according to the parabolic law, the exponent takes a value of $n = 0.25$. Here, the parabolic law means that the grain size in the intermetallic layer is proportional to the square root of the annealing time. According to the result of diffusion couple AU, $n$ is equal to 0.36 at $T = 473$ K and monotonically increases with decreasing annealing temperature. At $T = 393$ K, $n$ becomes close to 0.5. Even if the reactive diffusion is purely controlled by the grain boundary diffusion, $n$ is equal to 0.5 unless the grain growth occurs. Thus, for diffusion couple AU, the grain boundary diffusion contributes to the rate-controlling process, and the grain growth takes place at a certain rate at $T = 473$ K. However, at $T = 393$ K, the grain growth will slow down, though the contribution of the grain boundary diffusion becomes more remarkable. As can be seen in Fig. 7, the open squares coincide well with the open rhombuses. This deduces that the same temperature dependence of the rate-controlling process as diffusion couple AU works also for diffusion couple AG. On the other hand, for the open triangles, $n$ is close to 0.5 at $T = 433 \pm 473$ K. This indicates that the volume diffusion is mostly the rate-controlling process for diffusion couple GU. Although the intermetallic layer mainly consists of Ag$_3$Sn for both diffusion couples AG and GU, the grain boundary diffusion considerably contributes to the rate-controlling process for diffusion couple AG but hardly for diffusion couple GU. This is the reason why the growth rate of the intermetallic layer is smaller for diffusion couple GU than for diffusion couple AG. In contrast, for diffusion couple AP, the exponent takes values around $n = 0.35$ at $T = 433 \pm 473$ K. Furthermore, there is no systematic temperature dependence of the exponent. This means that the grain boundary diffusion appreciably contributes to the rate-
controlling process. Actually, the very fine two-phase microstructure is realized in the intermetallic layer for diffusion couple AP at \( T = 433\text{–}473 \text{ K} \). In such a two-phase microstructure, the effective cross-section for the grain boundary diffusion is fairly large. As a result, the growth rate of the intermetallic layer becomes greater for diffusion couple AP than for diffusion couple AG.

The logarithm of the proportionality coefficient \( k \) for diffusion couple AP is plotted as open circles against the reciprocal of the annealing temperature \( T \) in Fig. 8. As can be seen, the open circles are located rather well on a straight line. Thus, \( k \) is expressed as a function of \( T \) by the following equation:48)

\[
k = k_0 \exp(-Q_k/RT).
\] (3)

Here, \( k_0 \) is the pre-exponential factor, \( Q_k \) is the activation enthalpy, and \( R \) is the gas constant. From the open circles in Fig. 8, \( k_0 \) and \( Q_k \) were evaluated by the least-squares method. The evaluated values are shown in Fig. 8. Using these values, the temperature dependence of \( k \) was calculated from eq. (3). The result is shown as a solid line in Fig. 8. In this figure, the corresponding results of diffusion couples AU,17–19) AG,20) and GU23) are indicated as open rhombuses, squares and triangles, respectively. According to the results in Fig. 7, the almost same temperature dependence of \( n \) is actualized in the intermetallic layer for diffusion couple AP at each annealing time. Thus, from eq. (2), \( k \) is also one order of magnitude greater for diffusion couple AU than for diffusion couple AG. As shown in Fig. 6, however, \( l \) is one order of magnitude greater for diffusion couple AP than for diffusion couple AG. In contrast, \( l \) is slightly smaller for diffusion couple GU than for diffusion couple AG, but \( n \) is greater for diffusion couple GU than for diffusion couple AG. Consequently, \( k \) becomes smaller for diffusion couple GU than for diffusion couple AG.

As mentioned earlier, the two-phase microstructure consisting of the Ag\(_5\)Sn matrix dispersed with the very fine PtSn\(_4\) particle is realized in the intermetallic layer for diffusion couple AP at \( T = 433\text{–}473 \text{ K} \). In such a case, the grain boundary diffusion remarkably contributes to the rate-controlling process. The growth rate of the intermetallic layer is predominantly determined by the interdiffusion across the intermetallic layer.42,43) The temperature dependence of the interdiffusion coefficient \( D \) for the intermetallic layer is described by the following equation of the same formula as eq. (3) with certain accuracy.44)

\[
D = D_0 \exp(-Q/RT)
\] (4)

Here, \( D_0 \) is the pre-exponential factor, and \( Q \) is the activation enthalpy. Since the volume diffusion is the rate-controlling process for diffusion couple GU, we may expect that \( Q \) is close to 48 kJ/mol. However, there exists a rather complicated relationship between \( Q \) and \( Q_k \).44–48) Thus, the value of \( Q \) cannot be readily estimated from that of \( Q_k \) in a straightforward manner, even if the volume diffusion is the rate-controlling process.
across the Ag$_3$Sn matrix in the intermetallic layer for diffusion couple AP, Pt is the most minor component. Unfortunately, however, no reliable information of the intrinsic diffusion of Pt is available for the grain boundary diffusion in Ag$_3$Sn. The melting temperature $T_m$ of pure component is 505, 1235, 1337 and 2042 K for Sn, Ag, Au and Pt, respectively. Thus, $T_m$ is much higher for Pt than for Sn, Ag and Au. For a component with a high melting temperature, the intrinsic diffusion coefficient may be small, but the activation enthalpy will be large. The mean composition of the intermetallic layer for diffusion couple AP is obtained to be $y_{Ag} = 0.626$, $y_{Pt} = 0.033$ and $y_{Sn} = 0.341$ from the intersection of the dashed and dashed-and-dotted lines in Fig. 3. Assuming equivalent molar volumes of Ag$_3$Sn and PtSn$_4$, the volume fraction $f$ of PtSn$_4$ in the Ag$_3$Sn + PtSn$_4$ two-phase microstructure is estimated by the equation

$$f = \frac{y_{Sn} - y_{Sn}^0}{y_{Sn}^0 - y_{Sn}^1},$$

(5)

where $y_{Sn}^0$ and $y_{Sn}^1$ are the mol fractions of Sn for Ag$_3$Sn and PtSn$_4$, respectively. Inserting $y_{Sn} = 0.341$, $y_{Sn}^0 = 0.25$ and $y_{Sn}^1 = 0.8$ into eq. (5), we obtain $f = 0.165$. The value $f = 0.165$ indicates that a rather large amount of PtSn$_4$ is included in the intermetallic layer for diffusion couple AP. Consequently, the interdiffusion in PtSn$_4$ as well as the grain boundary diffusion in Ag$_3$Sn contributes to the overall interdiffusion across the intermetallic layer. In a previous study,$^{37}$ Sn/Pt/Sn diffusion couples were isothermally annealed at $T = 473$ K for various times. The Sn/Pt/Sn diffusion couple is hereafter called diffusion couple PT. During annealing, an intermetallic layer of PtSn$_4$ is formed at each interface in diffusion couple PT. The growth rate of the intermetallic layer is slightly smaller for diffusion couple Pt than for diffusion couple AG. This implies that the interdiffusion coefficient $D$ is smaller for PtSn$_4$ than for Ag$_3$Sn and thus the activation enthalpy $Q$ is greater for PtSn$_4$ than for Ag$_3$Sn. Since the activation enthalpy is large for the interdiffusion in PtSn$_4$ and the grain boundary diffusion in Ag$_3$Sn, the value of $Q$ for the overall interdiffusion may be also large. Although there is a rather complicated relationship between $Q$ and $Q_k$, the large value of $Q$ will result in that of $Q_k$. This may be the reason why $Q_k$ is greater for diffusion couple AP than for diffusion couples AU, AG and GU in Fig. 8.

On the basis of the temperature dependence of $k$ in Fig. 8, the value of $k$ at $T = 393$ K was calculated from eq. (3). The calculation provides $k = 3.08 	imes 10^{-6}$, $7.22 	imes 10^{-9}$ and $2.18 	imes 10^{-9}$ m for diffusion couples AP, AG and GU, respectively. On the other hand, the mean value of $n = 0.366$ is obtained from the experimental values at $T = 433$–473 K for diffusion couple AP. Since no systematic temperature dependence of $n$ is observed, $n$ is assumed to be 0.366 at $T = 393$ K for diffusion couple AP. As to diffusion couples AG and GU, however, $n$ is almost equal to 0.5 at $T = 393$ K. Using such values of $k$ and $n$, $l$ was calculated as a function of $t$ at $T = 393$ K from eq. (2). The results of diffusion couples AP, AG and GU are shown as solid, dashed and dashed-and-dotted lines, respectively, in Fig. 9. In this figure, the ordinate and the abscissa indicate the logarithms of $l$ and $t$, respectively. In contrast, the dependence of $l$ on $t$ was experimentally observed at $T = 393$ K for diffusion couple AU.$^{18}$ The experimental result of diffusion couple AU is shown as open rhombuses with a dotted line in Fig. 9. The dotted line gives $k = 5.39 	imes 10^{-9}$ m and $n = 0.477$. As can be seen in Fig. 9, the growth rate of the intermetallic layer is greater for diffusion couple AU than for diffusion couple AG but smaller for diffusion couple GU than for diffusion couple AG. Unlike Fig. 6, however, the growth rate becomes much smaller for diffusion couple AP than for diffusion couples AU, AG and GU. Consequently, it is concluded that the addition of Pt with 5 at% into Ag accelerates the kinetics of the reactive diffusion between Ag and Sn at $T = 433$–473 K but may decelerate the kinetics at $T = 393$ K. This means that the binary Ag–5Pt alloy is a superior conductor material resistant to the deterioration of the mechanical and electrical properties at the interconnection with the Sn-base solder alloy during energization heating unless the heating temperature is higher than 400 K.

4. Conclusions

The solid-state reactive diffusion in the ternary (Ag–Pt)/Sn system was experimentally examined using the Sn/Ag$_{90.05}$Pt$_{0.95}$/Sn, Sn/Au/Sn and Sn/Au$_{0.07}$Au$_{0.93}$/Sn diffusion couples are shown as solid, dashed and dashed-and-dotted lines, respectively. The experimental result of Sn/Au/Sn diffusion couple$^{18}$ is indicated as open rhombuses with a dotted line.
process for the growth of the intermetallic layer. The growth rate of the intermetallic layer is greater for the Sn/Ag$_{0.95}$Pt$_{0.05}$/Sn diffusion couple than for the Sn/Ag/Sn diffusion couple at $T = 433–473$ K. Considering the temperature dependence of the growth rate, however, the growth rate may become smaller for the Sn/Ag$_{0.95}$Pt$_{0.05}$/Sn diffusion couple than for the Sn/Ag/Sn diffusion couple at $T = 393$ K. Thus, the addition of Pt with 5 at% into Ag accelerates the kinetics of the reactive diffusion between Ag and Sn at $T = 433–473$ K but decelerates the kinetics at $T = 393$ K. This indicates that the binary Ag–5Pt alloy can be used as a conductor material resistant to the deterioration of the mechanical and electrical properties at the interconnection with the Sn-base solder alloy during energization heating unless the heating temperature exceeds 400 K.

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