The Influence of Radiation-Induced Vacancy on the Formation of Thin-Film of Compound Layer during a Reactive Diffusion Process

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Abstract. A theoretical approach is developed that describes the formation of a thin-film of *AB*-compound layer under the influence of radiation-induced vacancy. The *AB*-compound layer is formed as a result of a chemical reaction between the atomic species of *A* and *B* immiscible layers. The two layers are irradiated with a beam of energetic particles and this process leads to several vacant lattice sites creation in both layers due to the displacement of lattice atoms by irradiating particles. *A*- and *B*-atoms diffuse via these lattice sites by means of a vacancy mechanism in considerable amount to reaction interfaces *A/AB* and *AB/B*. The reaction interfaces increase in thickness as a result of chemical transformation between the diffusing species and surface atoms (near both layers). The compound layer formation occurs in two stages. The first stage begins as an interfacial reaction controlled process, and the second as a diffusion controlled process. The critical thickness and time are determined at a transition point between the two stages. The influence of radiation-induced vacancy on layer thickness, speed of growth, and reaction rate is investigated under irradiation within the framework of the model presented here. The result obtained shows that the layer thickness, speed of growth, and reaction rate increase strongly as the defect generation rate rises in the irradiated layers. It also shows the feasibility of producing a compound layer (especially in near-noble metal silicide considered in this study) at a temperature below their normal formation temperature under the influence of radiation.

Keywords: *Radiation-induced vacancy; Interfacial reaction; Diffusion; Critical thickness; Thin-film of AB-compound layer.*

I. INTRODUCTION

It is a well-established fact that irradiation creates defects in solid materials [1]. These defects are called radiation-induced defects and the simplest of them is termed a Frenkel pair (vacancy-interstitial pair). By means of radiation-induced defects many new channels are created for atomic diffusion in the target material. Atomic diffusion is very important for layer formation in solid-state reaction. The growth of a compound layer at the reaction interface requires both diffusion and a chemical reaction. This process is termed reactive diffusion.
The phenomenon of reactive diffusion is observed experimentally under both non-irradiation [2-5] and irradiation [6-9] conditions. The compound layer growth is shown to obey parabolic law (thickness is proportional to square root of time) in both cases.

In this paper the influence of radiation-induced vacancy is considered on the growth of the compound layer during a reactive diffusion process. Although both vacancy and interstitial defects are created at the same time during irradiation, we take into consideration the situation whereby the contribution of interstitial defect is separated from that of vacancy. With this approach, we are able to determine the contribution of vacancy defects to the growth of a compound layer. We begin by considering the chemical reaction rate as the approximation of product of reactants' atom number densities and assume that the growth rate of the compound layer is proportional to the reaction rate during both interfacial-reaction- and diffusion-controlled stages. Different defect generation rates are taken into account in each irradiated layer in order to determine how it influences the speed of growth, reaction rate, and thickness of a compound layer at a temperature below the compound layer formation under normal heat treatment processes such as conventional furnace and rapid thermal processing.

II. MODEL AND BASIC EQUATIONS

Suppose the irradiation of *A*- and *B*-layers leads to the creation of vacancy and interstitial atoms. Let us not consider a very high rate of defect generation; for example, which corresponds to reactor irradiation. In this case, the concentration of interstitial atoms is small, the correlation between fluxes of vacancy and interstitials is weak, and contributions of vacancy and interstitial mechanisms of diffusion can be considered separately. The vacancy mechanism is investigated in the present work.

The number of vacancies in the *AB*-layer increases as the radiation-induced vacancy adds up to the thermally generated ones. We assume that *A*- and *B*-atoms diffuse via a vacancy mechanism through the *AB*-compound layer to reaction interfaces *A/AB* and *AB/B* to form *AB*-compound layer. The thickness of the *AB*-layer formed owing to chemical reaction at interfaces *A/AB* and *AB/B* are designated by $g_a(t)$ and $g_b(t)$ respectively. The total thickness of the compound layer is given by $g(t) = g_a(t) + g_b(t)$. Let the *x*-axis be perpendicular to all the layers under consideration and $g(t) = 0$ before irradiation.

The diffusivity of *A*- and *B*-atoms through the vacancy mechanism under irradiation are $D_a^{v,irr}$ and $D_b^{v,irr}$ respectively, and the equation relating the diffusivity under irradiation to that of thermal diffusion is given by:

$$D_a^{v,irr} = \frac{n_a^v D_a^{v,th}}{n_a^{v,th}} \text{ and } D_b^{v,irr} = \frac{n_b^v D_b^{v,th}}{n_b^{v,th}}. \tag{1}$$

where n_a^v and n_b^v are vacancy concentrations under irradiation, $D_a^{v,th}$ and $D_b^{v,th}$ are diffusivities of *A*- and *B*-atoms via vacancy mechanism due to thermal vacancies, and $n_a^{v,th}$ and $n_b^{v,th}$ are thermal vacancy concentrations for *A*- and *B*-atom sites inside the *AB* layer.

Neglecting the recombination between vacancy and interstitial atoms, we can consider the change of densities of vacancy and interstitial atoms separately and obtain equations for two kinds of vacancy concentrations in the *AB* layer:

$$\frac{dn_a^v}{dt} = Kn_a^0 - \frac{\left(n_a^v - n_a^{v,th}\right)}{\tau_a^v} \text{ and } \frac{dn_b^v}{dt} = Kn_b^0 - \frac{\left(n_b^v - n_b^{v,th}\right)}{\tau_b^v}. \tag{2}$$

where K is the defect generation rate, n_a^0 and n_b^0 are the number densities of *A*- and *B*-lattice atoms in their respective layers, and $\tau_{v,}^a$ and τ_v^b are life-times for *A*- and *B*-vacancy. Because the relaxation of vacancy distribution occurs more quickly than the growth of the *AB*-layer, we assume that the vacancy distribution is homogeneous and stationary: $dn_a^v/dt = 0$ and $dn_b^v/dt = 0$.

Thus, for *A*- and *B*-vacancy concentrations we obtain:

$$n_a^v = Kn_a^0\tau_a^v + n_a^{v,th} \text{ and } n_b^v = Kn_b^0\tau_b^v + n_b^{v,th}. \tag{3}$$

Neglecting transient diffusion, we obtain equations for stationary concentration of *A* and *B* atoms inside the *AB* layer:

$$D_a^{v,irr}\frac{\partial^2 n_a(x)}{\partial x^2} = 0 \text{ and } D_b^{v,irr}\frac{\partial^2 n_b(x)}{\partial x^2} = 0 \tag{4}$$

with corresponding boundary conditions:

$$J_a\left(x = g_b\right) = R_a = v_a\gamma n_b^0 n_a\left(x = g_b\right); \quad J_b\left(x = -g_a\right) = R_b = v_b\gamma n_a^0 n_b\left(x = -g_a\right);$$

$$n_a\left(x = -g_a\right) = n_a^0; \quad n_b\left(x = g_b\right) = n_b^0 .$$

where J_a and J_b are the fluxes of *A*- and *B*-atoms, R_a and R_b represent the chemical reaction rate at *A/AB* and *AB/B* interfaces and they are not exactly the same, v_a and v_b are the stoichiometric coefficients of *A*- and *B*-atoms, and γ is the reaction rate constant. The fluxes of *A*- and *B*-atoms are assumed to occur at the same rate as the chemical reaction in order to ensure that the reaction only occurs at the interface and not inside either the *A*- or *B*-layer bulk.

The speed of growth of the *AB*-layer by virtue of chemical reaction at *A/AB* interface is $dg_a/dt = V_{ab}R_a = V_{ab}\gamma\, n_a^0 n_b$ (x = - g_a) and at *AB/B* is $dg_b/dt = V_{ab}R_b = V_{ab}\gamma\, (n_b^0)^2$, therefore the total speed is $dg/dt = dg_a/dt + dg_b/dt$, and V_{ab} is the volume of one molecule of the *AB*-layer.

By solving (4) with the given boundary conditions, we obtain an expression for the distribution of the *A*- and *B*-atoms inside the *AB*-compound layer:

$$n_a(x) = -\frac{\gamma\left(n_b^0\right)^2}{D_a^{v,irr}}\left(x + g_a\right) + n_a^0 \text{ and } n_b(x,t) = \frac{\gamma n_a^0 n_b^0 v_b}{D_b^{v,irr} + \gamma n_a^0 v_b g(t)}\left(x - g_b\right) + n_b^0. \tag{5}$$

III. THE GROWTH OF THE *AB*-LAYER

The rate of chemical reaction at interfaces of solid layers depends on the number densities of the *A* and *B* atoms in a complex manner. However, if there exists an excess of *A*- (or *B*-) atoms, then the rate of reaction remain the same with a change of the number density of *A*- (or *B*-) atoms [10]. On the other hand, if number densities of the *A*- and *B*-atoms approximately equal $v_a n_a \approx v_b n_b^0$, then the rate of the *AB*-compound formation can be represented in the first approximation as a product of *A*- and *B*-atom number densities.

Thus the rate of chemical reaction between the *A*- and *B*-atoms, at reaction interfaces *A/AB* and *AB/B*, can be approximately expressed in two stages in the following ways:

$$R_a = \begin{cases} \gamma\left(n_a^0\right)^2 = const & v_b n_b\left(x=-g_a\right) \geq v_a n_a^0 \\ \gamma n_a^0 n_b\left(x=-g_a\right) & v_b n_b\left(x=-g_a\right) \leq v_a n_a^0 \end{cases} .$$

$$R_b = \begin{cases} \gamma\left(n_b^0\right)^2 = const & v_a n_a\left(x=g_b\right) \geq v_b n_b^0 \\ \gamma n_b^0 n_a\left(x=g_b\right) & v_a n_a\left(x=g_b\right) \leq v_b n_b^0 \end{cases} . \tag{6}$$

Where n_b (x = - g_a) and n_a (x = - g_b) are number densities of the *B*-atoms at x = - g_a and the *A*-atoms at x = - g_b.

Owing to this, approximating the growth of the *AB* layer can have two stages. The first stage is when there is an excess of one kind of diffusing atom; for example, an excess of the *A*-atoms near interface *AB/B* or an excess of the *B*-atoms near interface *A/AB*. The second stage is when there is no excess of any kind of atoms near the corresponding surface; the second stage can take place independently of the first stage. However, if there is a first stage, the second stage must follow suit.

Thus, for the first stage, the growth rate of the *AB*-layer is determined by both the diffusion of the *A*- and *B*-atoms inside the *AB*-layer and also by the rate of reaction at the interface:

$$\frac{dg}{dt} = \frac{dg_a}{dt} + \frac{dg_b}{dt} = V_{ab}\left(R_a + R_b\right) = V_{ab}\left(\gamma\left(n_a^0\right)^2 + \gamma n_b^0 n_a\left(x = g_b\right)\right) . \tag{7}$$

The number density of the *A*-atom near interface *AB*/*B* and the *B*-atom near interface *A/AB* at second stage are:

$$n_a(g_b(t)) = \frac{D_a^{v,irr} n_a^0}{D_a^{v,irr} + \gamma n_b^0 v_a g(t)} \quad \text{and} \quad n_b(-g_a(t)) = \frac{D_b^{v,irr} n_b^0}{D_b^{v,irr} + \gamma n_a^0 v_b g(t)} . \tag{8}$$

If there is an excess of *A*-atom at the reaction interface, *AB/B* at time $t < t_c$, (t_c, is the critical time), the *AB*-layer at this interface grows under the interfacial-reaction-controlled process, and at the *A/AB* interface, the growth is controlled by diffusion; therefore the relationship between time and layer thickness is found by solving (7):

$$t = \frac{v_a g(t)}{\gamma v_b V_{ab}\left(n_b^0\right)^2} - \frac{D_b^{v,irr} v_a^2}{\gamma^2 v_b^3 V_{ab}\left(n_b^0\right)^3} In\left[\frac{\gamma n_a^0 n_b^0 v_b^2 g(t)}{D_b^{v,irr}\left(n_a^0 v_a + n_b^0 v_b\right)} + 1\right]. \tag{9}$$

However, if there is no excess of *A*-atoms at the reaction interface, no interfacial-reaction-controlled growth would occur at the *AB/B* interface, and the growth at both interfaces *A/AB* and *AB/B* would be predominantly diffusion controlled at all time.

At critical time t_c g(t) = g_c: $v_a n_a\left(g_b(t_c)\right) = v_b n_b^0$; where $n_a(g_b(t_c)) = \dfrac{D_a^{v,irr} n_a^0}{D_a^{v,irr} + \gamma n_b^0 v_a g_c}$. (10)

Critical thickness, g_c, is obtained as follows:

$$g_c = \frac{D_a^{v,irr}(n_a^0 v_a - n_b^0 v_b)}{\gamma v_a v_b \left(n_b^0\right)^2}. \tag{11a}$$

And for critical time substitute $t = t_c$, and $g = g_c$ in (9)

$$t_c = \frac{v_a g_c}{\gamma v_b V_{ab}\left(n_b^0\right)^2} - \frac{D_b^{v,irr} v_a^2}{\gamma^2 v_b^3 V_{ab}\left(n_b^0\right)^3} In\left[\frac{\gamma n_a^0 n_b^0 v_b^2 g_c}{D_b^{v,irr}\left(n_a^0 v_a + n_b^0 v_b\right)} + 1\right]. \tag{11b}$$

The growth rate of the compound layer at $t > t_c$ is described by:

$$\frac{dg(t)}{dt} = \gamma V_{ab}\left(n_a^0 n_b\left(-g_a(t)\right) + n_b^0 n_a\left(g_b(t)\right)\right). \tag{12}$$

By inserting (8) into (12) and integrating the resulting equation, a connection between time and layer thickness is established:

$$t = t_c + \alpha_1(g^2(t) - g_c^2) + \alpha_2(g(t) - g_c) - \alpha_3 \ln\left[\frac{\alpha_4 g(t) + \alpha_5}{\alpha_4 g_c + \alpha_5}\right]. \tag{13}$$

where:

$$\alpha_1 = \left[2V_{ab}\left[D_b^{v,irr} n_b^0 + D_a^{v,irr} n_a^0\right]\right]^{-1}, \alpha_2 = \left[\left(D_a^{v,irr} n_a^0\right)^2 + \left(D_b^{v,irr} n_b^0\right)^2\right]\left[\gamma V_{ab} n_a^0 n_b^0\left[D_b^{v,irr} n_b^0 + D_a^{v,irr} n_a^0\right]^2\right]^{-1},$$

$$\alpha_3 = D_a^{v,irr} D_b^{v,irr}\left[D_a^{v,irr} n_a^0 - D_b^{v,irr} n_b^0\right]^2\left[\gamma^2 V_{ab} n_a^0 n_b^0\left[D_b^{v,irr} n_b^0 + D_a^{v,irr} n_a^0\right]^3\right]^{-1}, \alpha_4 = \gamma\left[D_b^{v,irr} n_b^0 + D_a^{v,irr} n_a^0\right],$$

$$\alpha_5 = D_a^{v,irr} D_b^{v,irr}.$$

IV. RESULTS AND DISCUSSION

The growth kinetics of the thin-film of *AB*-compound layer expressed in (9) shows time as a linear and natural logarithmic function of thickness, and in (13) as a parabolic and natural logarithmic function of compound layer thickness. The natural logarithmic function in both (9) and (13) is ascribed to the simultaneous diffusion of both the *A* and *B*-atomic species via vacancy mechanisms during the compound layer growth.

If, for instance, only *A*-atoms diffuse in *AB*-layer when $D_b^{v,irr} = 0$ equation (9) becomes a linear equation:

$$g(t) = V_{ab}\gamma\left(n_b^0\right)^2\left(v_a\right)^{-1} v_b t. \tag{14}$$

and (13) reduces to parabolic equation:

$$\gamma n_b^0(g^2(t) - g_c^2) + 2D_a^{v,irr}\left(g(t) - g_c\right) - 2\gamma V_{ab} D_a^{v,irr} n_a^0 n_b^0\left(t - t_c\right) = 0. \tag{15}$$

and if only *B*-atoms diffuse when $D_a^{v,irr} = 0$, linear growth will be absent, only the parabolic law will hold equation (13), therefore, becomes:

$$\gamma n_a^0 g^2(t) + 2D_b^{v,irr} g(t) - 2\gamma V_{ab} D_b^{v,irr} n_a^0 n_b^0 t = 0\,. \quad (16)$$

Irradiation influences the growth of the *AB*-layer via change of diffusivity due to the production of the vacancy at rate *K*. Let us assume that the *A*-atoms diffuse only via vacancies of the *A*-sub-lattice of the *AB* compound and *B*-atoms via the *B*-sub-lattice, accordingly. This means that we consider diffusion of the *A*- and *B*-atoms separately and $D_{a(b)}^v = (D_{a(b)}^v = n_{a(b)}^v D^{v,a(b)})/n_{a(b)}$; taking into account that $(\tau_{v,a(b)}^{AB})^{-1} = (\rho d)_{a(b)} D_a^{v,a(b)}$ and $n_a^0 = v_a\, n_0$, $n_b^0 = v_b\, n_0$. Where $(\rho d)_a$ and $(\rho d)_b$ are the dislocation densities of *A*- and *B*-crystals, $D^{v,a}$ and $D^{v,b}$ are diffusivities of *A*-and *B*-vacancies, n_a and n_b are number densities of *A*- and *B*-atoms inside the *AB* layer. D_a^v and D_b^v are diffusivities of the *A*- and *B*-atoms via vacancy-mediated process; from (1) we obtain:

$$D_{a(b)}^{v,irr} = D_{a(b)}^{v,th} + \frac{K}{\rho_d v_{a(b)} n_0} \quad (17)$$

Where n_0 is the density of the *AB*-molecule. Thus, diffusivity is a linear function of the defect generation rate and we can estimate influence of irradiation using (9), (11), (13) - (16).

Table 1. Estimation of diffusivity of palladium and silicon in Pd_2Si layer under irradiation together with speed of growth and reaction rate of the layer.

$D_a^{v,irr}$ (Pd) $(10^{-20} m^2/s$	$D_b^{v,irr}$ (Si) $(10^{-16}\ m^2/s)$	dg_a/dt $(10^{-9}$ m/s)	dg_b/dt $(10^{-7} m/s)$	R_a $(10^{19}/m^2 s)$	R_b $(10^{21}/m^2 s)$	K (dpa/s)	γ $(10^{-36} m^4/s)$
0.047	0.021	0.290	0.120	0.690	0.275	10^{-9}	0.110
0.470	0.210	0.900	0.330	2.100	0.775	10^{-8}	0.310
4.700	2.100	3.200	1.200	7.600	2.750	10^{-7}	1.100

Table 2. Estimation of speed of growth and reaction rate of Ni_2Si layer with diffusivity of nickel and silicon in Ni_2Si layer under irradiation.

$D_a^{v,irr}$ (Ni) $(10^{-20} m^2/s$	$D_b^{v,irr}$ (Si) $(10^{-16}\ m^2/s)$	dg_a/dt $(10^{-9}$ m/s)	dg_b/dt $(10^{-8} m/s)$	R_a $(10^{19}/m^2 s)$	R_b $(10^{20}/m^2 s)$	K (dpa/s)	γ $(10^{-37} m^4/s)$
0.028	0.015	0.220	0.062	0.670	0.190	10^{-9}	0.075
0.280	0.150	0.770	0.190	2.300	0.575	10^{-8}	0.230
2.800	1.500	2.500	1.400	7.500	4.100	10^{-7}	1.640

Table 3. Estimation of diffusivity of platinum and silicon in Pt_2Si layer under irradiation and also the speed of growth and reaction rate of the layer.

$D_a^{v,irr}$ *(Pt)* $(10^{-20} m^2/s$	$D_b^{v,irr}$ *(Si)* $(10^{-16}\ m^2/s)$	dg_a/dt (10^{-9} m/s)	dg_b/dt $(10^{-7} m/s)$	R_a $(10^{19}/m^2s)$	R_b $(10^{21}/m^2s)$	K *(dpa/s)*	γ $(10^{-36} m^4/s)$
0.014	0.022	0.290	0.028	0.675	0.065	10^{-9}	0.026
0.140	0.220	0.960	0.240	2.200	0.550	10^{-8}	0.220
1.400	2.200	3.500	2.400	8.100	5.500	10^{-7}	2.200

Based on the model described above, we estimate the speed of growth, reaction rate, and layer thickness of three near-noble metal silicides at room temperature under irradiation with a defect generation rate of K = 10^{-9}, 10^{-8}, and 10^{-7}dpa/s. The reason for selecting this set of values was stated in the model and basic equation section. The thin-film of near-noble metal silicides is considered in this study because of their technological application and importance in integrated circuits [11].

The diffusivity under irradiation is estimated for the *A*- and *B*-atoms in A_2B-layers of palladium silicide, nickel silicide, and platinum silicide using (17) where *A*-atom represents near noble metal and *B*-atom denotes silicon. The result obtained from the diffusivity of both Pd- and Si- atoms as diffusing species under irradiation is shown in Table 1. The diffusivity of silicon is estimated to be about 10^4 times faster than that of palladium, which makes silicon the dominant species in the Pd_2Si layer. Similar results are shown in Tables 2 and 3 for nickel silicide, Ni_2Si, and platinum silicide, Pt_2Si . This result is contrary to the report in [12-14] where palladium, nickel, and platinum were reported as the dominant species under thermal diffusion. The same view holds for the other near-noble metal silicides where the metals were seen as the main diffusing species during the silicide growth [2, 12-13] in the first compound phase. Tables 5 and 6 contain data for the estimation of thermal self-diffusivity of Pd, Si, Ni, and Pt in their respective silicides. However, the growth of silicides under irradiation from our theoretical study indicate a view different from thermal diffusion in terms of the predominant diffusing species during the silicide formation, especially for the three silicides considered in this work. The estimated values of diffusivity of silicon in the three silicide layers under different defect generation rates at room temperature yielded results that lie within the range of integrated inter-diffusion coefficients of these silicides at their formation temperatures [2-5, 15] and this also strongly corroborates silicon as the active diffusing species in the silicide layer under irradiation. The formation temperature of these silicides are shown in Table 4.

The results depicted in Tables 1, 2, and 3 show that the speed of growth and reaction rate increase as the defect creation rate rises in the irradiated layer. This is as a result of opening of several channels (i.e. creation of vacant sites) in the irradiated layers for a considerable amount of atoms to diffuse to the reaction interface. The result also shows that the reaction in each layer proceeds at a different rate; likewise the layer growth speed. This difference can be attributed to the number of surface atoms present at the reaction interface during the layer formation. The interface that has more surface atoms has a greater chance of producing the compound layer faster than one with fewer atoms. The layer thickness depicted in Figures 1, 2, and 3 shows that the growth of the compound layer depends strongly on the defect generation rate. For instance, the temperature considered in this study is too low for silicide formation to take place in the absence of radiation. In the literature it has been reported that silicide formation can occur at a temperature much lower than room temperature (-120 °C) [16]. The tendency of producing silicides at a temperature as low as room temperature is highly feasible since radiation-enhanced diffusion depends strongly on a defect generation rate at such a low irradiation temperature.

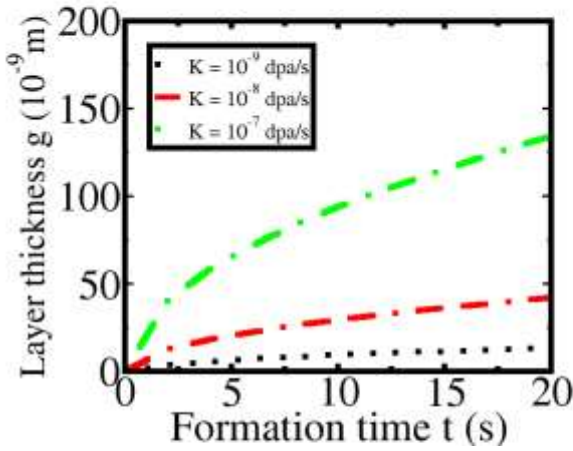

Figure 1: The growth kinetics of palladium silicide Pd_2Si at room temperature under the influence of radiation-induced vacancy at defect generation rate of K = 10^{-9}, 10^{-8}, and 10^{-7}dpa/s.

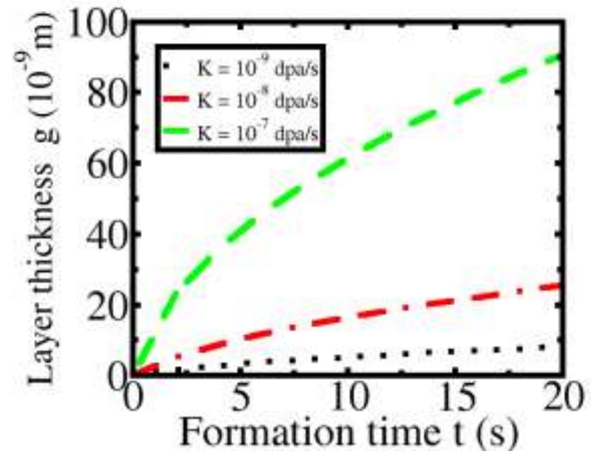

Figure 2: The growth kinetics of nickel silicide Ni_2Si at room temperature under the influence of radiation-induced vacancy at defect generation rate of K = 10^{-9}, 10^{-8}, and 10^{-7}dpa/s.

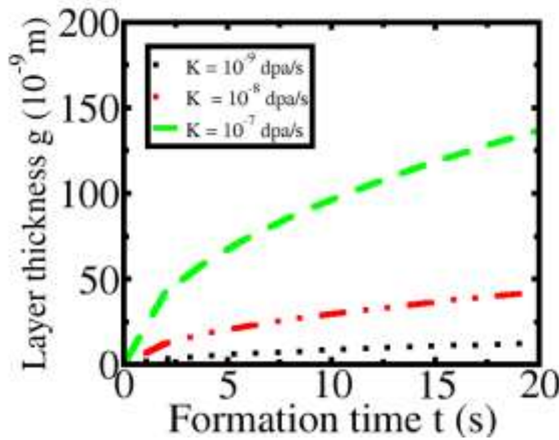

Figure 3: The growth kinetics of platinum silicide Pt_2Si at room temperature under the influence of radiation-induced vacancy at defect generation rate of K = 10^{-9}, 10^{-8}, and 10^{-7}dpa/s.

Table 4. Near-noble metal silicide formation temperature under non-irradiation condition (culled from [12]).

Compound layer	Temperature (K)
Pd_2Si	373
Ni_2Si	473
Pt_2Si	473

The growth kinetics shown in Figures 1, 2, and 3 depict a parabolic dependence of layer thickness on time. This result can be explained in light of the first phase of near-noble metal silicide.

In metal-rich silicide, metal is the majority species while silicon is the minority atoms. According to the model presented here, if 'majority atoms' (metal in this case) are active (i.e. dominant) during the layer growth, a reaction-controlled process would occur for a certain period of time and change to diffusion controlled process after which the growth kinetics has transformed to parabolic. However, if the dominant atoms are the minority species, then the layer growth would begin and end as a parabolic growth under a diffusion-controlled process. In a similar manner, silicon is presented as the minority and dominant species in the first compound phase of these silicides. Therefore, only one stage of growth is plausible (i.e. the parabolic growth). The silicide thickness is estimated with (13)

(when $g_c = 0$ and $t_c = 0$) over a period of 20 seconds at different defect generation rates in the three silicide layers.

Table 5. Parameters used for estimation of thermal self-diffusion coefficients of Pd, Ni, Pt, and Si in their respective silicides.

Volume of Compound layer (10^{-29} m^3)	Diffusing species under non-irradiation condition	Diffusing species under irradiation condition	Pre-exponential factor D_0 (m^2/s) for non-irradiation condition for atomic species	Activation energy E_a (eV) for non- irradiation condition for atomic species
Pd_2Si, (4.2)	Pd, Si [2]	Pd, Si [17]	Pd (1.5 x 10^{-8}) [14], Si (6.24 x 10^{-5}) [14]	Pd (1.0) [14], Si (1.7) [14]
Ni_2Si, (3.2)	Ni [12]	Ni, Si present work	Ni (1.82 x 10^{-3}) [2], Si (8.95 x 10^{-9}) [18]	Ni (1.71) [2], Si (1.9) [18]
Pt_2Si, (4.3)	Pt [13]	Pt, Si present work	Pt (5.5 x 10^{-4}) [19], Si (3.59 x 10^{-5}) [18]	Pt (1.485) [19], Si (2.1) [18]

Table 6. Parameters used for estimation of diffusion coefficient under irradiation at temperature of 298K.

Atomic Species	Dislocation density (10^9 m^{-2})	Vacancy formation energy for atomic species (eV)	Estimated value for number density of *AB*-molecule (10^{28} molecule/m^3)	Number density of atomic species (10^{28} atoms/m^3)
Pd	3000 [20]	1.70 [24]	Pd_2Si, (2.4)	Pd, (6.80)
Ni	5000 [21]	1.55 [25]	Ni_2Si, (3.3)	Ni, (9.14)
Pt	< 1000 [22]	1.35 [26]	Pt_2Si, (2.3)	Pt (6.50)
Si	1.0 [23]	2.32 [27]		Si (5.00)

IV. CONCLUSION

The results that follow from this study indicate that a thin-film of *AB*-compound layer is feasible under radiation-enhanced diffusion (radiation-induced vacancy type) for an *AB*-compound layer at a temperature lower than the formation temperature. The reason for this feasibility can be attributed to the dependence of diffusivity of atom, via vacancy mechanism, on defect generation rate at a low irradiation temperature. At such a low temperature, irradiation creates a number of vacant lattice sites in the *A*- and *B*- solid layers via which a substantial amount of *A*- and *B*- atoms are transported to the reaction interfaces, where they chemically react with surface atomic species that are close to the interfaces, thereby forming an *AB*-compound. The influence of room temperature on atomic diffusivity via vacancy is insignificant and cannot possibly enhance any growth at the reaction interfaces.

The results of this study also indicate that time dependence of layer thickness, i.e. growth kinetics, has both linear and natural logarithmic functions during the interfacial-reaction-controlled stage and parabolic and natural logarithmic functions during the diffusion-controlled stage. The reason for this result is ascribed to two forms of vacancy mechanisms in the *AB*-compound layer; one is an *A*-vacancy mechanism for the transport of *A*-atoms, and the other is the *B*-vacancy mechanism for transport of *B*-atoms. Lastly, we are able to show that defect generation rate enhances speed of growth, reaction rate and thickness of compound layer at low irradiation temperature.

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