



Formation and Growth Kinetic of Intermetallic Layers on 304SS at Different Slurry Aluminizing Temperature and Time Using Si-Al₂O₃-Modified Aluminide Coating

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Received: 1 May 2024 / Accepted: 15 March 2025
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Abstract The study explored how the presence of silicon (Si) and alumina influences the microstructures and growth kinetics of intermetallic compounds within aluminide coatings applied to 304 stainless steels. The aluminide coating was prepared via slurry aluminizing techniques and heat treated at 700 °C, 800 °C and 850 °C for 6, 8 and 10 h. The aluminide coating was characterized using field emission electron microscope-energy dispersive X-ray, X-ray diffraction and hardness. Multilayer phases were formed which comprise of binary phases (Fe₂Al₅, FeAl₃, FeAl and Fe₃Al), and the Fe–Si–Al ternary phases (Fe₂Al₃Si₃, Fe₄Al_{17.5}Si_{1.5} and Fe₄Al_{1.7}Si). As the aluminizing temperature and time increase, the thickness of the brittle intermetallic layer (Fe₂Al₅) layer decreases, while the thickness of the ductile intermetallic layers (FeAl and Fe₃Al) increases. The growth rate of the layers obeyed a parabolic law. The activation energy for Fe₂Al₅ was – 174 kJ/mol, while the activation energy for FeAl and Fe₃Al were 72 kJ/mol and 155 kJ/mol respectively. Microhardness measurement revealed that Fe₂Al₅, FeAl, and Fe₃Al layers had hardness values of 880–1010 HV, 570–630 HV, and 320–410 HV respectively at all aluminizing temperature and time. Fe₂Al₅ has the lowest toughness and the hardest zone in the aluminide coating. The Si addition effectively reduced the coating thickness of the

brittle intermetallic layers and aided the growth of ductile intermetallic layers.

Keywords Slurry aluminizing · Si-Al₂O₃-modified aluminide coating · Intermetallic layer · 304SS · Growth kinetics · Hardness

1 Introduction

Stainless steel and its alloys have been extensively used in hot corrosion and high-temperature oxidation. However, the prolonged use of these materials in hot corrosion environment degraded the corrosion resistance [1]. Surface modification has been used to improve the properties of these materials by depositing a suitable coating material on the surface called aluminide coating. Aluminizing iron-based alloys, nickel-based alloy or low carbon steel significantly increases corrosion resistance by promoting the formation of alumina scale in the oxidizing atmospheres. Among the aluminizing techniques are pack cementation (chemical vapour deposition), hot dip aluminizing (HDA) and slurry processes. The slurry technique has become increasingly popular due to its simplicity, cost-effectiveness, environmentally friendly process, and flexibility [2–4]. This method involves suspending the aluminium particles in a mixture of a solvent (either water or organic) and a binder, along with additional additives to maintain the stability and rheology of the slurry. The application of this slurry on metal surfaces can be performed through spraying, brushing, or immersion. Subsequent heat treatment in a furnace is used to cure the binder and create intermetallic layers [5].

Aluminide coating consists of several intermetallic compounds such as Fe₃Al, FeAl, FeAl₂, Fe₂Al₅, and FeAl₃. The formation of various Fe_xAl_y types is influenced by

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the composition of the aluminizing agent, as well as by the temperature and time of the aluminizing process [6]. Cheng and Wang [7] reported that the brittle intermetallic compound formed during aluminizing composed of FeAl_3 and Fe_2Al_5 with a tongue-like morphology serving as a stress concentrator that might initiate cracks during high temperatures. Addition of alloying elements such as silicon (Si) can influence the diffusivity of Fe and Al, affecting the formation and growth of the intermetallic layer in the aluminide coating. Hence, silicon (Si) plays an important role in the reactions and intermetallic phases during the aluminizing process which inhibits the formation of intermetallic layers by occupying the vacancy site of the c-axis of the crystal structure of Fe_2Al_5 , thus impeding the growth rate of Fe_2Al_5 layers and increase the activation energy required for the formation of the Fe_2Al_5 phase [8]. However, the effect of silicon does not only change the composition of the reaction layer but also has a pronounced, temperature-dependent effect on its growth kinetics by reducing the thickness of the intermetallic layer (IM) as compared to the conventional aluminide coating.

Growth kinetic is an important parameter during the formation of IMC which is a diffusion control for keeping the coating thickness at a minimal during the interaction between the coating and the substrate [9]. Aside from the growth kinetics, composition also has significant impacts on the growth rate of IMC layers [10]. Yin et al. [11] evaluated and found that the activation energies of Fe_2Al_5 phases were 207, 186, 169, 168, 167, and 172 kJ/mol for the growth rate at different silicon content (0–3.0%). Due to the parabolic nature of the activation energy of IMC layers, the result shows that the higher the silicon content, the lower the activation energy, where silicon occupied the vacancy site of the c-axis of Fe_2Al_5 crystal which have effect on its growth kinetics.

Researchers have observed that values of activation energies of IMC formed fall into various categories. Some activation energy values fall under 100 kJ/mol [12, 13], others reported in the range of 100–200 kJ/mol [14, 15] and above 200 kJ/mol [11, 16]. The discrepancies in activation energy values are attributed to the substantial impact of microstructure, surface condition, and other alloying elements. Although researchers have examined the growth kinetics of intermetallic layers, there remain notable gaps in understanding, particularly regarding the influence of alloying elements in aluminide coating on intermetallic layer interface growth kinetics, as only a limited number of studies have reported on this aspect. Thus, this research aims to investigate the formation and growth kinetic of intermetallic layer with Si- Al_2O_3 -modified aluminide coating on 304SS via slurry aluminizing. The investigation also includes coating thickness, morphology, and hardness of aluminide coating.

2 Experimental Procedures

2.1 Materials and Method

The substrate used is austenitic steel of grade 304SS. The material was cut into dimensions 10 mm × 10 mm × 2 mm which was then mechanically ground with 180 grit of silica carbide paper under flowing water to make a rough surface for quality slurry adhesion. Then the grease and oxides were removed from the surface of the specimen by ultrasonically cleaning in an acetone solution. Then the specimen was dipped in a NaOH solution and pickled in a solution of HCl and washed in distilled water. The specimen was then allowed to dry for 30 min.

The slurry was prepared by percentage mixing ratio of 57% solvent consisting of distilled water and Polyvinyl Alcohol (PVA) 14,000 as binder and 43% additives [17]. The additives are composed of a percentage mixing ratio of the powders (70% Al, 15% Al_2O_3 , and 15% Si). Alumina (Al_2O_3) functioned as an inert filler during the slurry aluminizing process. It absorbs additional heat, minimizing local melting, slowing the diffusion of aluminium, and influencing the growth kinetics of intermetallic compounds.

The prepared slurry was sprayed on the surfaces and edges of the sample with the aid of Sagola gun attached to the air pressure. The coating procedure was controlled to achieve a mass/area of 15–30 mg [18, 19]. The coated samples were then left to dry in laboratory air for 1 h. Coated samples were heat treated in a carbolite tube furnace under the argon environment at 750 °C, 800 °C and 850 °C for 6, 8 and 10 h to undergo an aluminizing process.

2.2 Coating Characterization

Cross section of the aluminide samples were mounted in an epoxy resin, ground with various silica carbide paper in the grade 240–2000 grits sizes and finally polished into a mirror like with alumina paste up to 0.05 µm and etched with a standard Keller etchant reagent for 3–5 s to reveal the microstructures of the samples. Surface morphology, cross-sectional microstructure and chemical composition were analysed by a FESEM (ZEISS SUPRA 35VP) with Energy Dispersive X-ray Spectrometer (EDX). X-ray diffraction (XRD Bruker Advanced D8) with Cu K α radiation ($\lambda = 0.15406$ nm) and 2θ scanning from 10° to 100° was performed on the samples for phases identification. Microhardness was performed on the coating's samples under the load of 25 g for 15 s. The coating layers thickness of the aluminized steels were examined using image J software on FESEM image.

2.3 Growth Kinetic Evaluation

The kinetics of aluminide layer formation was calculated using a parabolic law [10] as stated in Eq. 1.

$$X^2 = Kt, \quad (1)$$

where X (μm) stands for the coating thickness, K ($\mu\text{m}^2/\text{s}$) represents the growth rate constant, and t (s) is the aluminizing time of slurry. The Arrhenius equation expresses the relationship between the growth rate constant (K) and process temperature as shown in Eq. 2.

$$K = K_o \exp\left(\frac{-Q}{RT}\right) \quad (2)$$

K_o refers to the pre-exponential constant, Q represents activation energy (J), $T(K)$ indicates absolute temperature, and R is the gas constant 8.314 J/(K mol) .

From the growth rate constant, the activation energy (Q) of the aluminide coating prepared by different temperature and time are determined using Eq. 3.

$$\ln K = (\ln K_o) - Q/RT \quad (3)$$

3 Results and Discussion

3.1 Surface Image

Figure 1a shows a surface morphology representing the heat treated sample. During the aluminizing, there is interdiffusion between the coating (Al, Si, Al_2O_3) and the substrate resulting into the formation of intermetallic layers. The image revealed the coating uniformly distributed on the substrates. The formation of alumina scales on the aluminized stainless steel served as the protection oxides scale. While Fig. 1b shows a cross-sectional of the aluminide coating layer which consists of 2 layers; diffusion zone (DZ) and interdiffusion zone (IDZ) layer. The DZ comprises of aluminium (Al) and aluminium oxide (Al_2O_3), commonly

referred to as the topcoat. The IDZ serves as the transition region between the coating (Al, Si) and the substrate, where intermetallic formed.

3.2 SEM and EDX Analysis of the Aluminide Coating

Figure 2 shows the microstructure of the cross section of the aluminide coating on 304SS at 750°C , 800°C , and 850°C for 6, 8, and 10 h. The images revealed two dark grey intermetallic layers structures, i) outer layers, and ii) inner layers and the substrate. During the aluminizing process, the layers are continuous but vary in thickness along the interface. Additionally, the morphology of the coating towards the adjacent base materials changes with aluminizing temperature and time. Intermetallic layer formation in aluminide coatings occurs through diffusion-driven phase transformations at high temperatures. During heat treatment, aluminium diffuses into the substrate material, reacting with base metal atoms such as iron to form intermetallic compounds like Fe–Al binary. The process typically involves initial solid-state diffusion, followed by nucleation and growth of intermetallic phases. The layer structure evolves through a sequence of phase transformations, with Al-rich phases formed near the surface and Fe-rich phases formed near the substrate. The presence of silicon in aluminide coatings modifies diffusion kinetics, forming stabilized intermetallic phases of Fe–Al–Si ternary phases. The Fe–Al–Si phases were identified according to morphology, EDX analysis and previous research [20–23]. The outer layer consists of Al-rich zone (Fe_2Al_5 , FeAl_3), have low toughness and the inner layer consists of Fe-rich zone (FeAl and Fe_3Al) which are ductile binary phases, and the ternary phases ($\text{Fe}_4\text{Al}_{1.7}\text{Si}$, $\text{Fe}_2\text{Al}_3\text{Si}_3$ and $\text{Fe}_4\text{Al}_{17.5}\text{Si}_{1.5}$) segregated out in the intermetallic (IM) layers [24].

Pores were detected within the intermetallic layers due to the Kirkendall effect, where the different diffusion rates of Fe and Al lead to a net flux of vacancies, accumulating into voids within the protruding, isolated serrated steel. With prolonged exposure to high temperatures during aluminizing, adjacent voids merge, increasing their size, interconnected

Fig. 1 a Surface morphology of the aluminide coating, and b cross section image of the aluminide coating

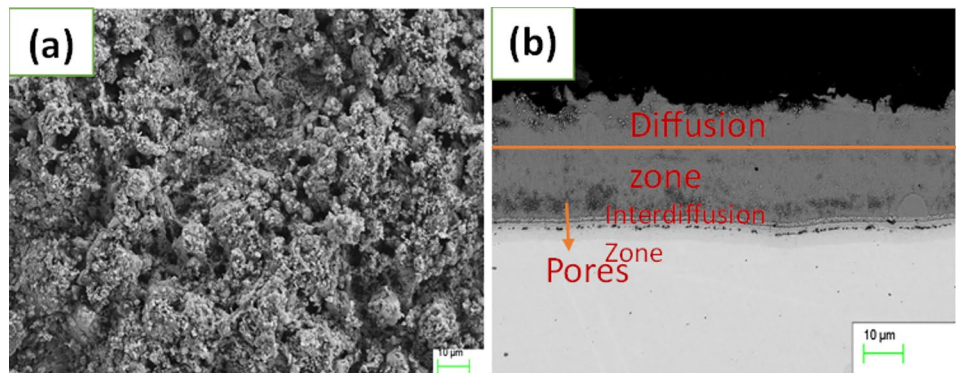
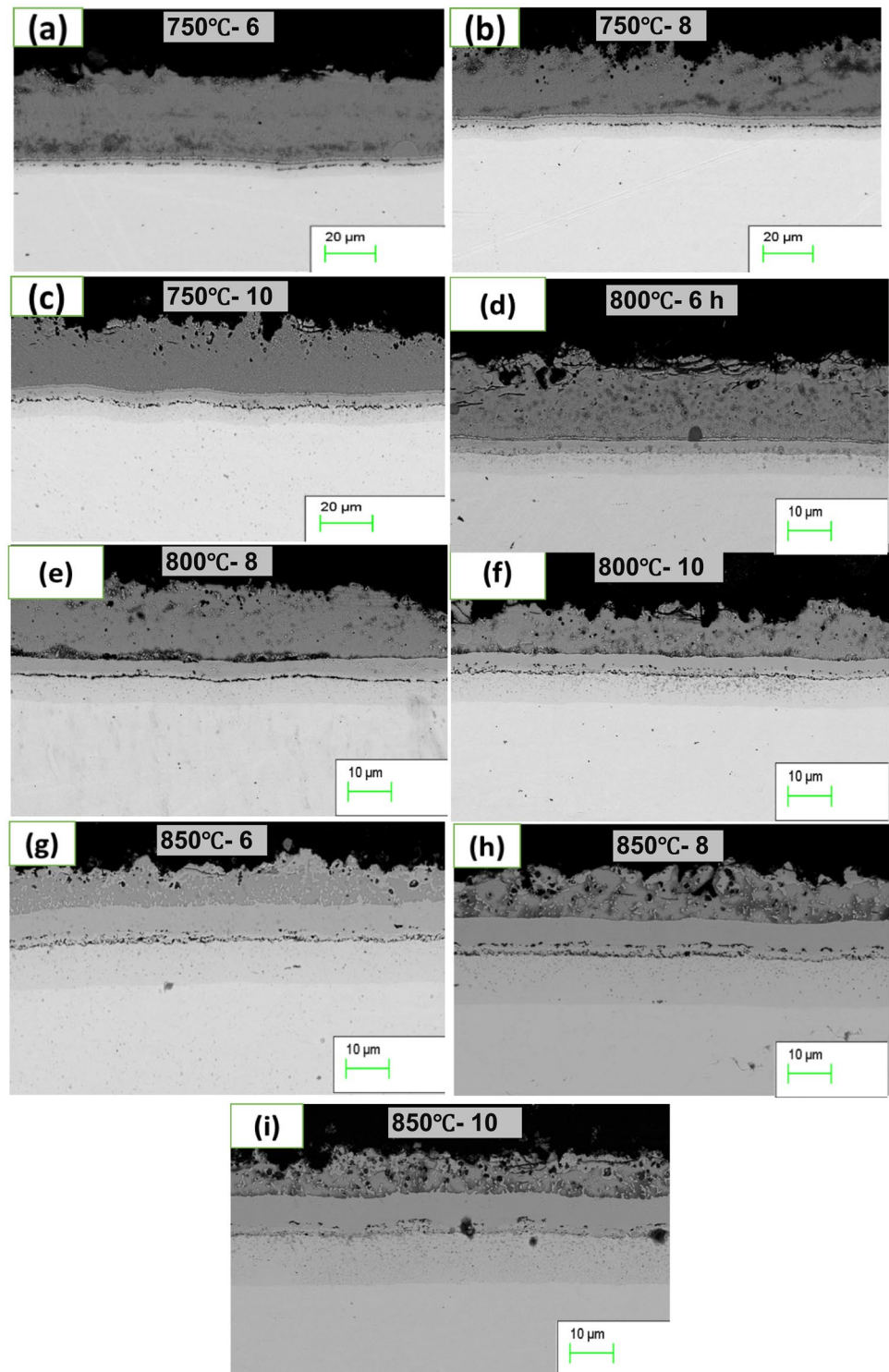


Fig. 2 FESEM micrographs of cross-sectional images of Si-Al₂O₃-modified aluminide coating. **a–c** 750 °C, **d–f** 800 °C, and **g–i** 850 °C for 6, 8 and 10 h



and forming pores. In addition, pores also formed due to uneven volume shrinkage during the transition from Fe₂Al₅ to FeAl. The increase in the number of pores is attributed to the increase in aluminizing time and temperature [6, 25]. The effect of pores was more pronounced at higher temperatures of 800 °C and 850 °C which may have negative impact

on the mechanical properties of the coating. Pores in aluminide coatings can deteriorate mechanical properties by acting as stress concentrators, reducing strength, toughness, and adhesion, while promoting crack initiation and propagation.

Table 1 shows the EDX analysis, which quantifies the chemical compositions of the aluminide layers at

Table 1 EDX data analysis on cross section of 304SS after aluminizing at 800 °C for 6 h

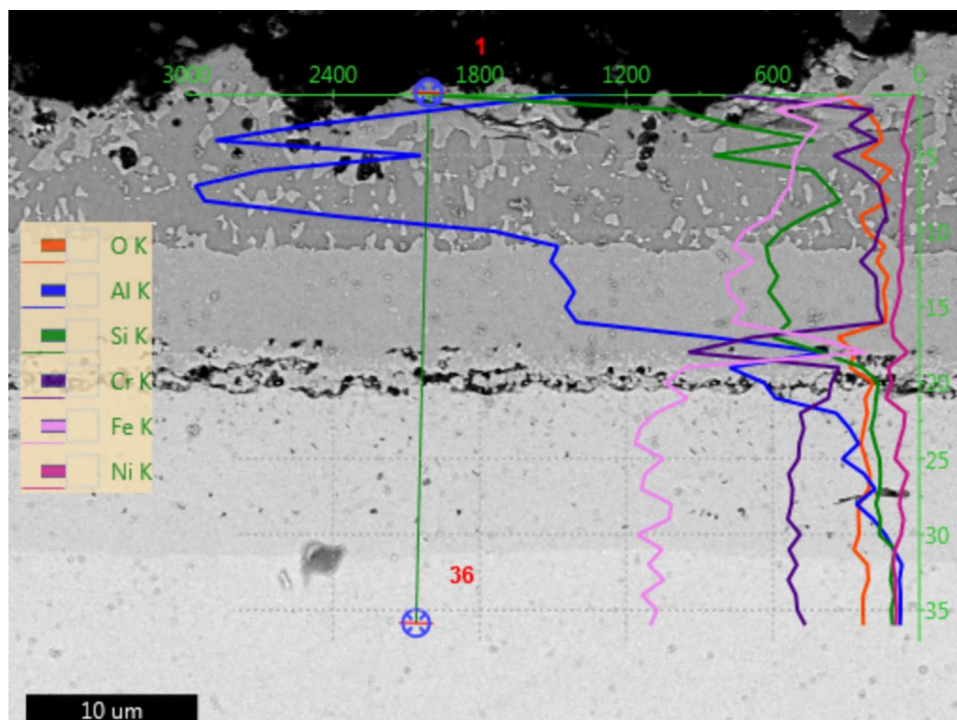
Spots	Composition (at.%) for 800 °C at 6 h						Phases
	Fe	Al	Si	Cr	O		
1	71.29	1.38	2.1	22.55	2.68		Substrate
2	65.72	7.1	4.29	19.31	3.58		Fe ₃ Al
3	38.53	38.57	14.65	5.2	3.06		FeAl
4	14.58	44.82	10.92	10.18	19.5		Fe ₂ Al ₃ Si ₃
5	22.94	53.32	21.16	1.4	1.9		Fe _{1.7} Al ₄ Si
6	32.47	58.2	12.48	3.94	6.22		Fe ₂ Al ₅

various points within the coating layers. This quantification is based on the percentage ratio of Al, Fe, and Si by atomic percent (% at.) for sample heat treated at 800 for 6 h. EDX analysis revealed that at point 6, the Fe–Al intermetallic primarily comprises major phases of Fe₂Al₅, whereas close to the substrate, at points 2 and 3, the predominant phases are FeAl and Fe₃Al, respectively, recognized as Fe-rich intermetallic. The sequence of formation for Fe–Al intermetallic compounds follows Fe₂Al₅ > FeAl₃ > FeAl > Fe₃Al. The initial formation of the Fe₂Al₅ phase occurs due to its lowest Gibbs free energy. Conversely, the free energy of the Fe₃Al phase exceeds zero, preventing its formation at temperatures higher than 400 °C. The introduction of Si results in the formation of the ternary phase Fe–Al–Si (Fe₄Al_{1.7}Si) that is consistent with findings from previous studies [14, 21, 26], while the emergence of Fe₂Al₃Si₃ represents a novel phase, as shown in Table 1.

EDX line scan in Fig. 3 shows the concentrations of each element involved in the formation of the intermetallic phases with aluminium having the highest at the topcoat layer and decreasing down to the substrates. In contrast, the concentration of Fe is high near the substrate. As seen from the EDX line scanning results, some changes in element percentage occur at the interdiffusion layer.

3.3 Phase Analysis

XRD analysis was used to validate the presence of intermetallic layer phases, as shown in Fig. 4. The detected phases include Fe₂Al₅ with (ICDD PDF number: 00-029-0043), FeAl (ICDD PDF number: 04-013-5555) and Fe₃Al (ICDD PDF number: 04-013-9780), all of which are primary phases. FeAl₃ was not detected, possibly due to either weak peak or its size being too small for detection by XRD. The ternary phases identified include: Fe_{1.7}Al₄Si (ICDD PDF number:

Fig. 3 EDX line scan of Si–Al₂O₃-modified aluminide coating on 304SS at 850 °C for 6 h

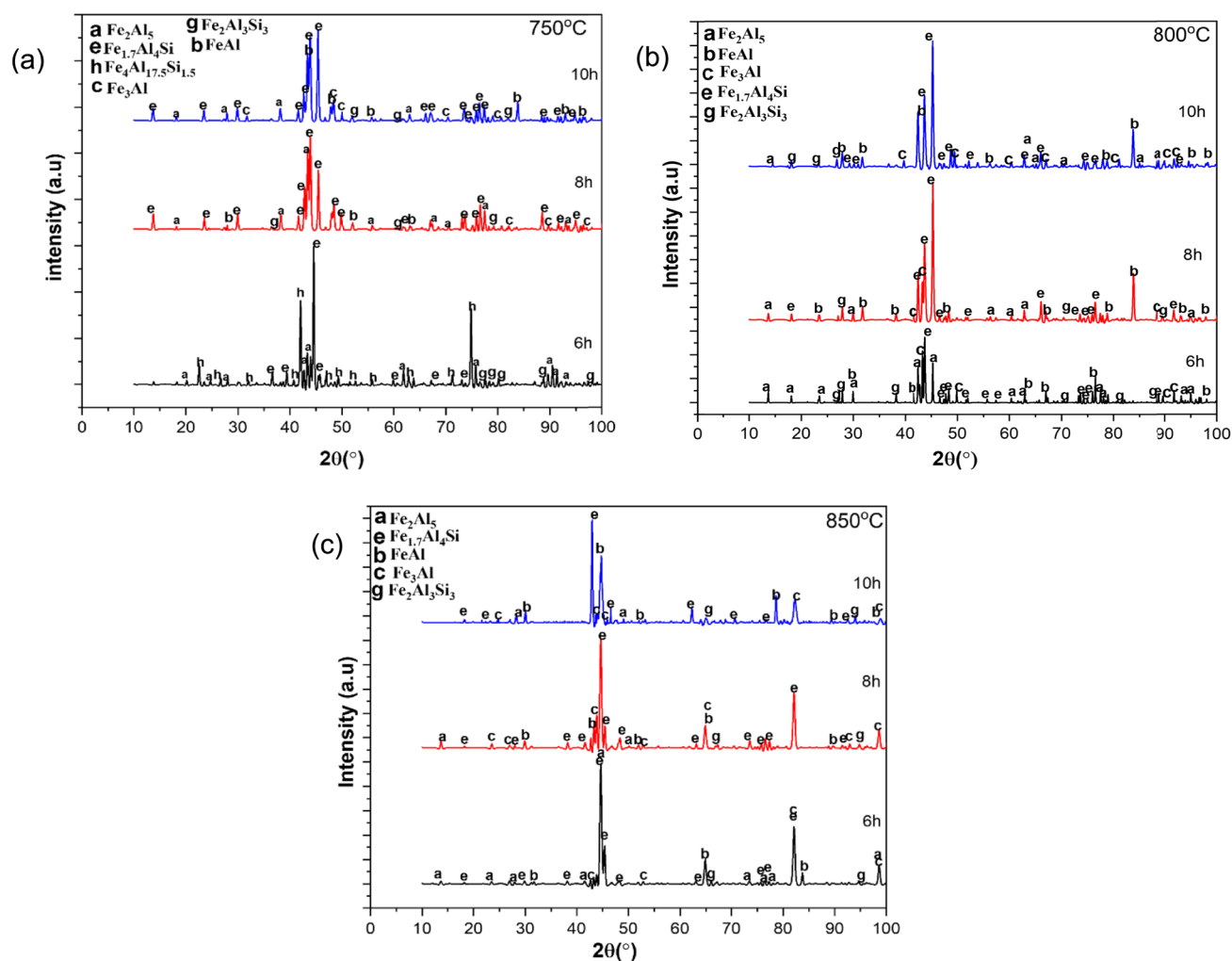


Fig. 4 XRD pattern of aluminized 304SS heat treated at **a** 750 °C, **b** 800 °C and **c** 850 °C for 6, 8 and 10 h

04-008-4784), $\text{Fe}_4\text{Al}_{17.5}\text{Si}_{1.5}$ (ICDD PDF number: 04-003-1389) and $\text{Fe}_2\text{Al}_3\text{Si}_3$ (ICDD PDF number: 01-984-4861). These findings are consistent with the literatures [21, 24, 27] except the unstable phase $\text{Fe}_4\text{Al}_{17.5}\text{Si}_{1.5}$ which was found at 750 °C, 6 h. Microstructural observations, together with EDX and XRD analyses indicate the presence of Fe_2Al_5 , FeAl , Fe_3Al , $\text{Fe}_2\text{Al}_3\text{Si}_3$, $\text{Fe}_4\text{Al}_{17.5}\text{Si}_{1.5}$, and $\text{Fe}_{1.7}\text{Al}_4\text{Si}$ phases at the interface of the aluminide coating on 304SS.

3.4 Microhardness

Figure 5 illustrates the hardness values of the $\text{Si-Al}_2\text{O}_3$ -modified aluminide coating at temperatures of 750 °C, 800 °C, and 850 °C across all soaking times. The hardness of Fe_2Al_5 falls within the range of 880–1010 HV, consistent with reported values ranging from 800 to 1030 HV in the literature [16, 28] thus demonstrating good agreement with the previous studies. Hardness values decrease progressively towards the substrates. FeAl exhibits hardness

values ranging from 570 to 630 HV, in line with reported values of 550–680 HV [18]. Similarly, the hardness of Fe_3Al recorded within the range of 320–410 HV. Among the samples, those treated at 850 °C for 10 h display the highest hardness value of 1010 HV for Fe_2Al_5 , while the lowest value occurs at 750 °C for 6 h. The increase in hardness is attributed to prolonged heat treatment, temperature, and the addition of alloying [28, 29]. The hardness increases with prolonged heat treatment, temperature, and alloying additions. Prolonged soaking time and higher temperatures enhance aluminium diffusion, leading to increased coating thickness and the formation of intermetallic compounds. The addition of silicon modifies diffusion kinetics by promoting the formation of hard intermetallic phases such as $\text{Fe}_{1.7}\text{Al}_4\text{Si}$, and $\text{Fe}_2\text{Al}_3\text{Si}_3$. Additionally, Si helps suppress the excessive formation of the brittle Fe_2Al_5 intermetallic phase, improving the overall coating properties. Consequently, these factors enhance the hardness of the coating.

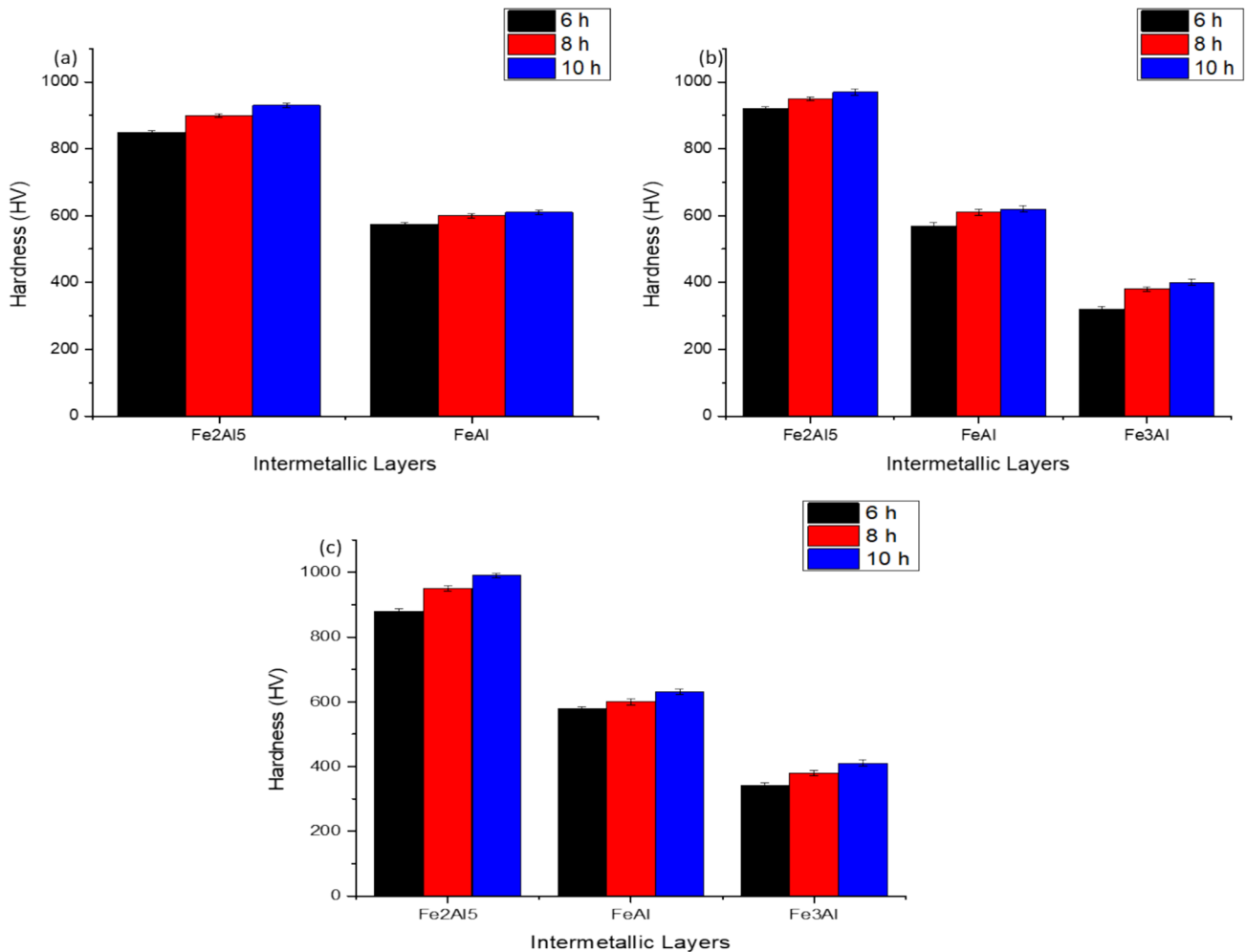


Fig. 5 Hardness of Si-Al₂O₃-modified aluminide coating layers on 304SS at **a** 750 °C, **b** 800 °C and **c** 850 °C for 6 h, 8 h and 10 h

Fe₃Al displays the lowest hardness, followed by FeAl. These layers are responsible for the corrosion and oxidation resistance of the aluminide coating at high temperatures [30]. Hardness directly impacts the strength of the aluminide coating layer under cyclic thermal shock in high-temperature applications. The formation mechanism of aluminide coating during aluminizing is driven by the adsorption of active aluminium atoms diffusing toward the substrate to form intermetallic compounds [31, 32] as seen in Fig. 1b. The formation and growth kinetics of intermetallic layers are influenced by elemental composition, which determines diffusion rates, phase stability, and coating hardness. Heat treatment temperature impacts microstructure and phase transformations, accelerating diffusion and phase evolution. The incorporation of silicon plays a crucial role in reducing the brittleness of intermetallic compounds and enhancing coating hardness. The aluminium concentration in each layer determines the hardness and progressively decreases through diffusion toward the substrate, as shown in Fig. 3.

3.5 Growth Kinetics Result of Intermetallic Layer

Figure 6 shows the correlation between the square of the aluminide layers thickness between the diffusion time (6, 8 and 10 h) for all the aluminizing temperatures (750 °C, 800 °C and 850 °C). The thickness of each intermetallic layers formed during the aluminizing process were shown in Table 2. The square of thickness shows a linear relationship with the diffusion time, according to Eq. (1). This means that there is a parabolic relation between the thickness of the aluminide layers and time at various aluminizing temperatures. Interdiffusion between the coating and substrate determines the growth rate of the coating layers. The growth rate of intermetallic compounds such as Fe₂Al₅, FeAl, and FeAl₃ are controlled by diffusion rate. This has been proposed and reported in literatures [10, 16, 32]. All the intermetallic compounds (Fe₂Al₅, FeAl and Fe₃Al) followed the same trend as shown in Fig. 6. The slope of the plotted line in Fig. 6 determines the growth rate constant (*K*) values.

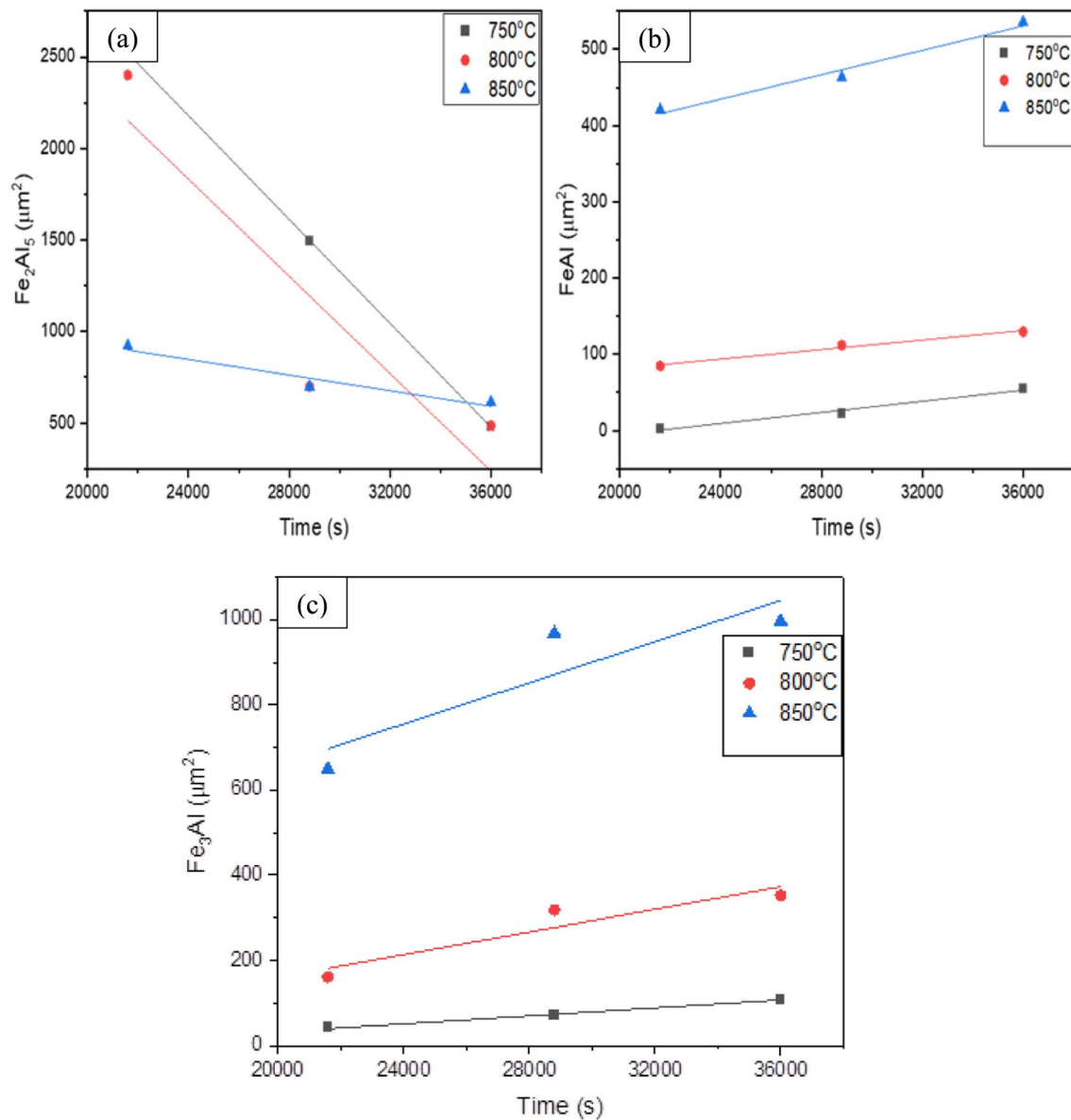


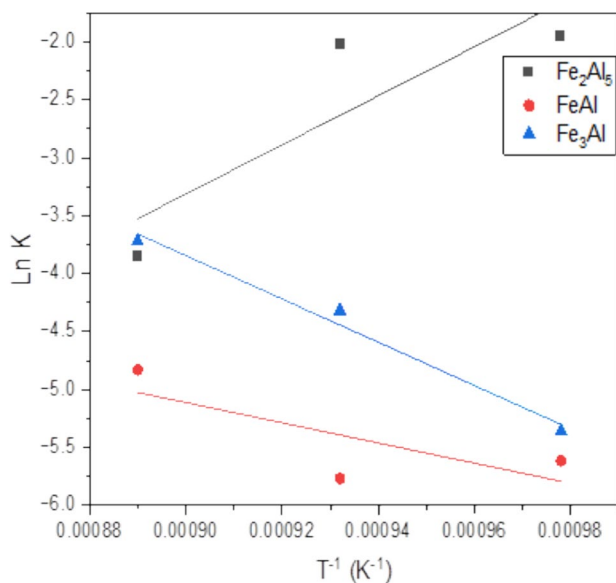
Fig. 6 Relationship between the square of intermetallic layer thickness and diffusion time **a** Fe_2Al_5 , **b** FeAl , and **c** Fe_3Al

Figure 7 depicts the $\ln(K)$ against the reciprocal of temperature ($1/T$). This figure indicates that there is a linear relationship between $\ln(K)$ and $1/T$. Equations (1) and (2) show that the slopes of plotted lines in Fig. 7 are equivalent to Q/R . The activation energy (Q) was determined from the slope Q/R obtained from the graph for each intermetallic layer as shown in Table 3. These obtained activation energies were then compared with the reference values as shown in Table 4. The results indicated that the reduction in Fe_2Al_5 growth rate due to reduction of thickness, was influenced not only by temperature and aluminizing time but also by silicon occupying the vacancy sites within Fe_2Al_5 crystal structure [8, 30, 33]. The presence

of silicon facilitated the formation of ductile phases such as FeAl and Fe_3Al . The pre-exponential factor and activation energy values for the intermetallic compounds are shown in Table 3. Fe_2Al_5 exhibits the lowest pre-exponential factor and an activation energy of -174 kJ/mol, whereas Fe_3Al displays the highest value. Consequently, the calculated activation energies for FeAl and Fe_3Al are 72 kJ/mol and 155 kJ/mol, respectively. This result falls within the previous findings [9, 16, 34–36], although the previous findings focused solely on Al and Fe reactions. The overall activated energy (Q) obtained for Al bath was 53 kJ/mol, while Q calculated for this study Al–Si was 43 kJ/mol conformed with the finding of [11].

Table 2 Average coating thickness (μm) of intermetallic layer after aluminizing process

Tempera- ture ($^{\circ}\text{C}$)	IMC	6 h	8 h	10 h
750	Fe_2Al_5	50.23 (± 4.35)	36.64 (± 4.16)	21.89 (± 6.22)
	FeAl	1.66 (± 0.29)	4.73 (± 0.63)	7.44 (± 0.96)
	Fe_3Al	6.44 (± 1.22)	8.38 (± 0.55)	10.47 (± 0.81)
800	Fe_2Al_5	49 (± 8.43)	26.43 (± 8.10)	22.04 (± 9.7)
	FeAl	9.20 (± 0.47)	10.58 (± 0.75)	21.52 (± 2.50)
	Fe_3Al	12.71 (± 1.24)	17.85 (± 0.84)	18.79 (± 1.14)
850	Fe_2Al_5	30.36 (± 5.18)	26.42 (± 5.3)	24.79 (± 7.21)
	FeAl	20.51 (± 1.14)	21.52 (± 2.50)	23.15 (± 2.93)
	Fe_3Al	25.47 (± 0.92)	31.57 (± 1.28)	31.57 (± 0.07)

**Fig. 7** Logarithmic relationship between growth rate constant (K) and $1/T$ **Table 3** Value of activation energy of the IMC from the slope of Fig. 7

IMC	($-Q/R$)	$\ln K_0(\mu\text{m/s})$	$Q(\text{KJ/mol})$	$K_0(\text{m/s})$
Fe_2Al_5	21,218.03745	-22.40932	-174	1.85×10^{-16}
FeAl	-8723.16689	2.73893	72	1.55×10^{-5}
Fe_3Al	-18,667.28049	12.95505	155	4.23×10^{-1}

* Q/R -Slope, Q -activation energy, $\ln K_0$ -Intercept, K_0 -Pre-exponential constant

Zou et al. [37] and Lemmens et al. [8] specifically investigated the impact of silicon addition on Fe_2Al_5 growth, confirming that activation energy decreases with increasing silicon content. The presence of ternary phases

Table 4 Comparison of activation energy, Q (KJ/mol) of different aluminide layers calculated and reported in literature

	Fe_2Al_5	FeAl	Fe_3Al
This work (calculated)	-174	72	155
Literature	74.1 [39]	180 [9]	
	100 [37]	249.52 [16]	260 [9]
	141 [9, 40]	340 [40]	
	203 [16]		
	-72 [8]		

can potentially influence the growth of the Fe_2Al_5 phase, either by nucleation or by serving as a diffusion barrier. In this study, the ternary phases $\text{Fe}_{1.7}\text{Al}_4\text{Si}$ and $\text{Fe}_2\text{Al}_3\text{Si}_3$ were detected within the intermetallic compound due to silicon addition. Silicon enrichments at grain and phase boundaries of the Fe_2Al_5 phases were observed as nano-scaled ternary phases, as previously reported, which act as diffusion barriers, leading to reduced diffusion rates [8].

The growth rate refers to the rate at which a specific variable or quantity increases within a defined timeframe. The presence of silicon in substantial quantities had notable effects on the formation, diffusion, and potential enhancement of certain ternary phases. The negative activation energy of Fe_2Al_5 suggests that the growth of the Fe_2Al_5 phase might have been impeded during the reaction between the coating and substrates to accommodate the formation of more ductile phases. The addition of Si encourages the formation of ductile phase (Fe_3Al and FeAl) within the aluminide coating, a critical aspect for enhancing corrosion resistance [20, 38]. The Fe_3Al and FeAl intermetallic phases enhance corrosion resistance primarily due to their ability to encourage the formation of stable and protective aluminium oxide (Al_2O_3) layer on the surface. This oxide layer acts as a barrier, reducing the diffusion of corrosive species such as oxygen and chloride ions. Additionally, these phases exhibit strong atomic bonding and low defect densities, making them less susceptible to localized corrosion mechanisms like pitting and intergranular attack. Their high thermodynamic stability further contributes to corrosion resistance.

4 Conclusions

The Si- Al_2O_3 -modified aluminide coatings were successfully developed on the 304SS through the slurry technique at 750 $^{\circ}\text{C}$, 800 $^{\circ}\text{C}$, and 850 $^{\circ}\text{C}$ for 6, 8 and 10 h. The morphology, elemental composition, hardness, phase analysis and growth kinetics were critically examined. Therefore, it could be concluded that:

- 1) A multi-layered coating structure was observed after the aluminizing for all coating samples. The intermetallic layers formed comprised of binary phases (Fe_2Al_5 , FeAl) and Fe_3Al and the ternary phases ($\text{Fe}_{1.7}\text{Al}_4\text{Si}$, $\text{Fe}_2\text{Al}_3\text{Fe}_3$ and $\text{Fe}_4\text{Al}_{17.5}\text{Si}_{1.5}$). The $\text{Fe}_4\text{Al}_{17.5}\text{Si}_{1.5}$ was the new phase formed in this study.
- 2) A temperature of 750 °C was chosen due to the presence of Fe–Al and Fe–Al–Si phases in the heat treated sample, which exhibited a dense structure without porosity. In contrast, pores were observed in samples treated at 800 °C and 850 °C, potentially compromising the mechanical properties of the coating for high-temperature applications.
- 3) The hardness of the FeAl and Fe_3Al layers ranged from 550 to 680 HV and 320–410 HV, respectively, indicating low ductility. In contrast, the Fe_2Al_5 layer exhibited a hardness range of 880 to 1010 HV across all aluminizing temperatures and heat treatment durations. This increased hardness was attributed to the incorporation of silicon would enhance the coating performance.
- 4) The addition of silicon to the aluminide coating suppresses the growth of Fe_2Al_5 , while promoting the formation of stable Al–Fe–Si phases, aiding in the development of ductile phases such as FeAl and Fe_3Al that is beneficial for corrosion resistance.
- 5) The growth rate of the aluminide layer follows a parabolic law, where the layer thickness exhibits a linear relationship with both time and aluminizing temperature. This indicates that the growth of aluminide layers on 304SS using the slurry aluminizing technique is diffusion-controlled.
- 6) The overall activation energy for the aluminide layer was 43 kJ/mol. For the Si-modified coating on 304SS, the activation energy of Fe_2Al_5 was calculated as – 174 kJ/mol, while FeAl and Fe_3Al exhibited activation energy of 72 kJ/mol and 155 kJ/mol, respectively.

Acknowledgements The authors gratefully acknowledge the School of Materials and Mineral Resources Engineering, Universiti Sains Malaysia for facilities provided and funding provided by the Ministry of Higher Education Malaysia under Fundamental Research Grant Scheme (FRGS) with Project Code: FRGS/1/2021/TK0/USM/02/22 and the sponsorship by TETFund, Nigeria (TETF/ES/UNIV/KWARA/TSAS/ 2021).

Author Contributions Ibrahim Owolabi Ambali was involved in investigation, methodology, writing—original draft. Anasyida Abu Seman was involved in supervision, conceptualization, writing—review and editing, methodology, funding acquisition. Tuti Katrina Abdullah was involved in conceptualization, supervision. Brij Kumar Dhindaw was involved in reviewing and editing.

Funding Fundamental Research Grant Scheme (FRGS), FRGS/1/2021/TK0/USM/02/22, Anasyida Abu Seman.

Declarations

Conflict of interest The authors declare that they have no conflict of interest.

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