Microstructure-based modeling of IMC layer thickness effect on tensile behavior of Cu/Al clad strip

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Keywords: clad strip, cold roll bonding, intermetallic compound, tensile property, microstructure-based modeling

Abstract

A microstructure-based model is proposed for quantitative estimation of tensile mechanical property of Cu/Al clad strips. In the proposed model, the fracture criterion for matrix Cu and Al, interface intermetallic compounds (IMCs) and interface between matrix and IMCs are established and incorporated. The effect of IMCs thickness and strip thickness on tensile behavior of Cu/Al strip was studied quantitatively using the proposed model. And a new dimensionless parameter \( w/t \) is also proposed to be a control index for tensile mechanical property of Cu/Al clad strips. And a critical value of \( w/t \) which is defined to be 0.025 in this study is found to be important in controlling mechanical property of Cu/Al strip with a thickness less than 0.5 mm. The potential application of the proposed model and parameter is also discussed for annealing process optimization and mechanical property control of 0.12 mm thick Cu/Al clad strips.

1. Introduction

In recent years, Cu/Al bimetals have been increasingly used in a variety of applications including electrical components and heat exchangers due to combined advantages of copper and aluminum [1]. Cold roll bonding is reported to be the most efficient method for large size flat clad metal sheets or strips, compared with the other processes such as explosion welding and diffusion bonding [2, 3]. For cold roll bonding of Cu/Al strips, an annealing treatment will be normally needed after rolling to increase the bond strength [2]. Several intermetallic compounds were found to form at the interface during annealing process. The formation and growth kinetics of IMCs during annealing process has been widely studied by both experimental and modeling methods [4–8]. The authors of this paper have just given a review of the modeling methods and proposed an inter diffusion based analytical model to investigate the growth kinetics of intermetallic compounds formed under various annealing and bonding conditions [9].

These brittle intermetallic compounds have also been found to significantly influence the mechanical properties and performance reliability of the Cu/Al sheets [3, 10–14]. For example, C Y Chen et al [3] studied the influence of interfacial structure development at interface on the fracture mechanism and the bond strength of Cu/Al bimetal plate. The main interfacial layers were determined to be Al₁₂Cu, Al₆Cu₅, Al₄Cu₉ and Al₆Cu₅ phases. The formation and thickening of those intermetallic compounds were found to promote cracks propagation and weaken the bond strength of the bimetal plates. The fracture mechanism transformed from ductile to brittle cleavage with the development of interfacial structures. N Y Kim and S I Hong [10] studied the interactive deformation of Cu/Al/Cu clad plates, and found that the interactive deformation of Cu and Al layers in layered Cu/Al/Cu with strong interface bonding has a beneficial effect on the ductility. However, In the presence of thick brittle interfacial intermetallic, no beneficial effect of mutual constraint on the enhanced deformability was attained and ductility even decreased because of interface cracking and separation during deformation. X B Li et al [11] studied the influence of IMC layer on tensile property of Cu/Al clad sheets, and revealed the interfacial layer with sub-micron thickness was helpful to the dislocation motion and improves the ductility of the Cu/Al...
Clad sheet, while the excessive formation of intermetallic compounds induced the interface delamination and reduced the tensile properties of the clad sheet. S Lee et al [14] studied the effect of bonding interface on delamination behavior of drawn Cu/Al bar clad material. It was evident from V-notch impact tests that the growth of the brittle diffusion layers with the increasing aging time after draw bonding directly influenced delamination distance between the Cu sleeve and the Al core.

Although the previous studies revealed some understanding on the fracture mechanism and qualitatively IMC layer dependence of the mechanical property of the Cu/Al clad sheets, the studies were mainly based on experimental results, quantitative dependence of IMC layer of the mechanical property is still unclear which is very important for improving or controlling the mechanical property due to lack of efficient modeling method of the deformation behavior of clad sheet considering the influence of interfacial layers.

Meanwhile, previous studies on roll bonded Cu/Al clad sheets were mainly focused on those with a thickness over 0.5 mm used for conducting bars. And less study had been carried out on thin Cu/Al clad strips which have great application potential for cable strips. An investigation on thin Cu/Al clad strip showed that the mechanical property was more sensitive to IMC layer compared to thicker clad sheets, and the mechanical property deteriorated quickly when the thickness decreased from 0.5 mm to 0.12 mm [15] which caused fracture failure of strips during performance. However the operating mechanism and controlling methods were still unclear. Thus it is very important to quantitatively describe the dependence of mechanical property on IMC layer, especially for thin strips, to control and improve the mechanical property for better performance.

The objectives of the present study are to establish a model based on real microstructure of thin Cu/Al clad strips including matrix and interfacial layers to investigate the quantitative influence of intermetallic compounds thickness on mechanical properties, and the application of the modeling results for controlling the mechanical property of a 0.12 mm thick clad strip is also presented. The study can improve the understanding of IMCs dependence of mechanical property and offer efficient method for quantitative assessment and process optimization for performance improvement for clad sheets.

2. Materials and methods

2.1. Materials preparation
Roll bonded Cu/Al strips with 0.35 mm and 0.12 mm thickness were used in the present study. Roll bonding experiments were carried out using annealed c11000 copper and 1050 Aluminum sheets. The roll bonding process is schematically illustrated in figure 1, and the detailed description of the process can be found in previous published paper [9]. The tensile specimen was then cut from the roll bonded 200 mm wide strips, the dimension of which is shown in figure 2. The annealing process was done in an electric furnace, and a thermal couple is welded on the specimen to monitor and control the heating temperature and holding time during annealing. Then Uniaxial tensile tests were performed on an Instron 5582 universal testing machine at room temperature, with an initial strain rate of 10^{-3}. The tensile stress-strain curve of roll bonded 0.35 mm thick Cu/Al strip after annealing under 793 K for 2 min holding is shown in figure 3.

The microstructures of the clad strip before and after tensile test were observed using SSX-550 scanning electron microscope (SEM), SHIMAJZU Wavelength-Dispersive Spectrometer (WDS) and Philips pw1050 x-ray diffraction technique (XRD) and the detailed description can be found in [9]. Three IMCs were identified in the interface including Al_{2}Cu, AlCu and Al_{4}Cu_{9} phases, and the thickness was measured to be 2.88 μm, 1.42 μm and 3.63 μm respectively for 0.35 mm thick strip [9]. And it is also found that the tensile fracture of IMCs layer occurred in AlCu and developed along AlCu and Al_{4}Cu_{9} phases, as shown in figure 4, which is also in consistent with the reported results [10–14] as AlCu phase is the mostly hard and brittle compounds compared to the other two.

2.2. The establishment of the microstructure-based model
Figure 5 shows the process used to obtain the microstructure-based model. Firstly, the geometrical model of the tensile specimen with Cu and Al layers according to figure 2 was built using an explicit finite element analysis software Abaqus, as shown in figure 3(a). Then the interfacial microstructure was identified with three intermetallic layers, as shown in figure 3(b), and the detailed identification process of the phase composition and width of each layer can be found in [9]. After that, the three interfacial layers can be added to the previous built geometrical model, as shown in figure 3(c). Finally, the geometrical model was meshed followed by loading and constrained boundary conditions. An object-oriented finite element software (OOF, National Institute of Standards and Technology, U.S. Department of Commerce) is used for meshing. The mesh is adapted to geometrical model by the OOF software, which allows the movement of the element nodes to conform to the model. Therefore, each element involves the properties of a single material (matrix or each interfacial layer). The model was meshed using the 4-node bi-linear plane strain quadrilateral (CPE4R in Abaqus). The mesh size is
adjusted to make sure at least one layer of mesh can be obtained for each material layer. The microstructure-based model was then used to predict the deformation behavior of the clad strip under uniaxial tensile load.

As the interfacial intermetallic phases $\theta$ (Al$_2$Cu), $\gamma_2$ (Al$_4$Cu$_9$), $\eta_2$ (AlCu) are all brittle compounds with small fracture strain, small deformation can induce fracture in the compounds; and the compounds can be a block for slip, which constrained the deformation of Cu and Al matrix and also caused easy debonding and fracture along interface between matrix and compounds. Therefore, the fracture of Cu/Al clad strips could include three kind of mechanisms: ductile fracture of matrix metals, brittle fracture of interfacial compounds and interface debonding between matrix and compounds. So, the models for different fracture mechanism are given below, which were also used to describe the fracture of particle reinforced metal matrix composites in previous study [16].

2.2.1. Damage model for the matrix metals
In clad bimetal sheets, the ductile fracture of the metal matrix can be modeled by damage theory proposed by Gurson for porous materials in which the Mises criteria are modified by introducing void volume fraction $f$ as a damage variable [15]. As the interaction between voids was not considered in Gurson model, there were found

Figure 1. Schematic illustration of roll bonding experimental process of Cu/Al clad strips [9].

Figure 2. Dimension of tensile specimen.
big deviation between modeling results and experiments. Then Tvergaard and Needleman later extended the model to the following form (the so called GTN model) [16]:

\[ \Phi = \frac{\sigma^2_m}{\sigma^2_m} + 2f^*q_1 \cosh\left(\frac{3q_1\sigma_{hk}}{2\sigma_m}\right) - (1 + q_3 f^*) = 0 \]  \hspace{1cm} (1)

In which, \( \Phi \) is the yielding function; \( \sigma_m \) is the yielding stress; \( \sigma_{eq} \) is the equivalent stress of the matrix material; \( \sigma_{hk} \) is the hydrostatic pressure; \( q_1, q_2 \), and \( q_3 \) are parameters reflecting interaction of voids; \( f^* \) is the rapid loss of stress carrying capacity due to void coalescence, which can be defined as follows:

\[
f^* = \begin{cases} 
  f & \text{if } f \leq f_c, \\
  \frac{f_c}{f_c} + \frac{f - f_c}{f_c - f_c} (f - f_c) & \text{if } f_c < f < f_F, \\
  \frac{f}{f_F} & \text{if } f \geq f_F,
\end{cases}
\]  \hspace{1cm} (2)

In which, \( f \) is the void volume fraction; \( f_c \) is the critical void volume fraction for occurrence of coalescence; \( f_F \) is the void volume fraction at failure; \( f_{bc} \) is the balance coefficient, which is defined as:
The GTN model defines failure through $f_{c}$ and $f_{p}$, for void coalescence when $f_{c} < f < f_{p}$ and materials failure when $f \geq f_{p}$.

The increase of void volume fraction comes from two parts: growth of original voids and newly formed voids, namely:

$$ \dot{f} = \dot{f}_{gr} + \dot{f}_{nucl} $$

Where, $\dot{f}_{gr}$ is the growth rate for original voids; $\dot{f}_{nucl}$ is the growth rate of void volume fraction due to newly formed voids. The two growth rates can be defined as:

$$ \dot{f}_{gr} = (1 - f) \frac{\varepsilon_{pl}^{m}}{\varepsilon_{N}} $$

$$ \dot{f}_{nucl} = A \frac{\varepsilon_{pl}^{m}}{\varepsilon_{N}^{2}} $$

$$ A = \frac{f_{N}}{\varepsilon_{N}^{2} \sqrt{2\pi}} \exp \left[ -\frac{1}{2} \left( \frac{\varepsilon_{pl}^{m} - \varepsilon_{N}}{\varepsilon_{N}} \right)^{2} \right] $$

In which, $f_{N}$ stands for volume fraction of particles which caused voids nucleation; $\varepsilon_{N}$ and $\varepsilon_{N}$ are mean values and standard deviation of normal distribution function; $\varepsilon_{pl}^{m}$ is the equivalent strain; $\varepsilon_{N}^{2}$ is the equivalent strain rate.

2.2.2. Damage model for the brittle IMC layers

In this study, the normal stress criterion [17] is employed to detect crack initiation of the brittle IMC layers:

$$ \max(\sigma_{1}, \sigma_{2}, \sigma_{3}) = \sigma_{0} $$

where $\sigma_{1}$, $\sigma_{2}$, $\sigma_{3}$ are the principle stress components, and $\sigma_{0}$ is the tensile strength of the IMCs. when the maximum tensile stress exceeded the tensile strength, cracks began to appear in the material.

The crack propagation behavior is described by the fracture energy method [18]. It takes the fracture energy of type I crack per unit area as a material parameter in brittle fracture. In this way, the brittle fracture behavior of brittle materials is transformed from the stress-strain response to the stress-displacement response.

The relationship between fracture displacement, fracture energy and fracture stress are defined as:

$$ u_{mb} = 2G_{f} / \sigma_{tu}^{l} $$

Where $G_{f}$ is suggested to be $40 \sim 120 \text{ N m}^{-1}$ [19].

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**Figure 5.** The process for construct the microstructure-based model (a) geometrical model for tensile specimen with Cu/Al matrix layer; (b) interfacial microstructure identification; (c) adding interfacial layers on the tensile specimen; (d) meshing and boundary condition of the specimen with both matrix and interfacial layers.
The evolution mechanism of type II crack is considered in the crack evolution process after the initiation of crack, that is to say, the shear modulus of the material also decreases in the process of crack opening. Therefore, it is necessary to define the relationship between shear modulus and crack opening displacement:

\[ G_c = \rho(e_{ck}^{\infty})G \]  

(10)

In which, \( G \) is the shear modulus before cracking; \( G_c \) is the shear modulus after cracking; \( \rho(e_{ck}^{\infty}) \) is the shear retention factor and is a function of crack opening strain \( e_{ck}^{\infty} \).

Specific expressions of \( \rho(e_{ck}^{\infty}) \) can be defined as follows:

\[ \rho(e_{ck}^{\infty}) = \left(1 - \frac{e_{ck}^{\infty}}{e_{ck}^{\max}}\right)^\rho \]  

(11)

Where, \( \rho, e_{ck}^{\max} \) are material constants. Generally, the tensile fracture strain is in the range of \( 10^{-4} \) to \( 10^{-3} \).

2.2.3. Damage model for the debonding of the interface between matrix and IMCs

Interface is an important factor affecting load transfer, stress distribution and fracture behavior in clad metals. The strength of interface bonding largely determines the macroscopic mechanical properties of clad metals. The bonding modes of the interfacial IMC layers and matrix in clad metals include physical bonding, mechanical bonding, chemical reaction bonding and diffusion bonding. There is an optimal binding strength of the interfacial IMC layers and matrix in clad metals include physical bonding, mechanical bonding, chemical reaction bonding and diffusion bonding. There is an optimal binding strength and is a function of crack opening strain \( e_{ck}^{\infty} \). In damage fracture analysis, the mechanical properties of the interfacial layer without considering the thickness and microstructure of the interface layer can be studied. Therefore, the interface damage can be describe by the cohesive behavior.

The linear elastic constitutive equation can be applied for interface material:

\[
t = \begin{bmatrix} t_n \\ t_s \\ t_t \end{bmatrix} = \begin{bmatrix} K_{nn} & K_{ns} & K_{nt} \\ K_{sn} & K_{ss} & K_{st} \\ K_{tn} & K_{ts} & K_{tt} \end{bmatrix} \begin{bmatrix} \delta_n \\ \delta_s \\ \delta_t \end{bmatrix} = K\delta
\]  

(12)

Where, \( t \) is the nominal traction stress tensor; \( K \) is the stiffness coefficient, and was set to 108 GPa which is large enough to ensure the continuity of displacement at interfaces[16]; \( \delta \) is the interfacial crack opening displacement.

In this study, the maximum stress criterion was used to determine the generation of interface damage. When the maximum contact stress coefficient reaches 1, the interface damage begins. The maximum stress criterion can be expressed as follows:

\[
\max\left\{ t_n, t_s, t_t \right\} = 1
\]  

(13)

Where, \( t_n \) is the interface normal contact stress; \( t_s \) is the contact stress in the first shear direction of the interface; \( t_t \) is the contact stress in the second shear direction of the interface; \( (\cdot) \) stands for the Macaulay bracket, which means:

\[
t_n = \begin{cases} 0, & t_n < 0 \\ t_n, & t_n \geq 0 \end{cases}
\]  

(14)

In which, \( t_n^0, t_s^0, t_t^0 \) are respectively the maximum contact stress in the corresponding direction.

The interface damage evolution process starts after the interface reaches the initial damage condition. The interface damage evolution equation describes the process of the interface stiffness reduction. The damage variable \( D \) is used to describe the damage on the contact surface, and the initial value is 0. When the interface reaches the initial condition of damage, the value of the damage variable \( D \) starts to change from 0 to 1. The relationship between stress component and damage variable can be defined as [17]:

\[
t_n = \begin{cases} t_n^0, & t_n \geq 0 \\ (1 - D)\tilde{t}_n, & \text{otherwise} \end{cases}
\]  

(15)

\[
t_s = (1 - D)\tilde{t}_s \]  

(16)

\[
t_t = (1 - D)\tilde{t}_t \]  

(17)

\( \tilde{t}_n, \tilde{t}_s, \tilde{t}_t \) are corresponding stress components on interface without damage.

In order to describe the damage evolution behavior of the interface under the combined action of normal stress and shear stress, Camanho and Davila [20] introduced the concept of equivalent separation displacement,
which is expressed as follows:

\[ \delta_m = \sqrt{\delta_m^2 + \delta_l^2 + \delta_l^2} \]  

(18)

linear softening of variable \( D \) is adopted for the interface damage evolution:

\[ D = \frac{\delta_f (\delta_m^{\text{max}} - \delta_m^0)}{\delta_m^{\text{max}} (\delta_f - \delta_m^0)} \]  

(19)

Where, \( \delta_m^0 \) is the equivalent separation distance at the beginning of interface damage; \( \delta_m^{\text{max}} \) is the equivalent separation distance at interface fracture, which was set to 1 \( \mu \)m according to [21]; \( \delta_m^{\text{max}} \) is the maximum value of the equivalent separation distance in the loading history.

### 2.3. Parameter identification for the microstructure-based model

For the identification of damage model parameters, there are usually several methods including experimental curve calibration method [22], optical analysis method [23] and numerical calculation method [24]. The experimental curve calibration method is used in this study.

As shown in figure 4 and the reported results in [10–14], the fracture of IMCs occurred in AlCu and Al\(_2\)Cu layer. So, for the brittle fracture model for IMCs, all the parameters are identified according to the property of AlCu and Al\(_2\)Cu phases. As there are few results available for the mechanical property of AlCu and Al\(_2\)Cu phases, the Young’s modulus of 172.21 GPa calculated of AlCu phase [25] and Poisson’s ratio of 0.29 calculated for Al\(_2\)Cu phase [26] are used for elastic parameters of IMCs layer. The tensile fracture strength of 6.44 GPa and fracture strain of 0.086 calculated for Al\(_2\)Cu phase [27] is used for brittle fracture parameters of IMCs layer.

For the GTN damage model of matrix Cu and Al layers, as parameters \( q_1, q_2, q_3 \) are insensitive to materials and loading conditions, the values of 1.5, 1.0, 2.25 suggested by Tvergaard [16] are used.

The other parameters are obtained from a uniaxial tensile test through best-fit methods. Firstly, a microstructure-based finite element model for the tensile test experiment of Cu/Al strip is established following the procedure describe in figure 5 to conduct a virtual tensile test. The numerical stress-strain curves and experimental one was compared after each numerical simulation completed. The simulated tensile curve is consistent with the experimental curve after repeated simulation and modification of parameters. The final identified dimensionless parameters was shown in table 1. Figures 6 and 7 show the comparison between the tensile specimen and stress-strain curve calculated by Finite element model using identified parameters and the experimental results. It can be seen that the calculated results are in good agreement with the experimental ones, which can justify the identified model parameters.

### 3. Modeling results

#### 3.1. Fracture evolution of 0.35 mm thick Cu/Al clad strip with different IMC layer thickness

Figure 8 shows the distribution of equivalent stress at different moments during the uniaxial tension process of the 0.35 mm thick Cu/Al clad strip with totally 7.93 \( \mu \)m thick IMCs layer simulated by the proposed microstructure-based finite element model. It can be seen from the figure that at the beginning of tension, the equivalent stress in copper layer is relatively larger. With the continue of tension, the stress in IMCs layer increases gradually, especially after necking of Al layer, which produces large stress concentration in the IMCs layer. Then fracture occurs at the interface between IMCs lay and Al layer (seen as the arrow indicated in figure 8(c)) and extends to the whole IMCs layer (seen as the arrow indicated in figure 8(d)), followed by ductile fracture of Al layer and finally Cu layer, resulting in tensile failure of the whole specimen.

Due to larger yield stress of Cu layer than Al layer, stress concentration occurs in the Cu layer at the beginning of tension. And thus, stress difference between Cu and Al layer causes quick necking of Al layer. With tension continue, as the interfacial brittle IMCs layer can’t coordinate deformation, stress concentration begins to produce in the IMCs layer, and the necking of Al layer intensifies this stress concentration, leading to crack at the interface of IMCs and Al layer. So it implies that uncoordinated deformation of IMCs layer is the root cause for tensile failure of Cu/Al strip. Therefore, the IMCs layer thickness should be an important factor for tensile property of Cu/Al strip.

### Table 1. The identified dimensionless parameters for GTN damage model of matrix Cu and Al layers.

| Layer | \( q_1 \) | \( q_2 \) | \( q_3 \) | \( f_0 \) | \( f_1 \) | \( f_N \) | \( \epsilon_N \) | \( s_N \) |
|-------|--------|--------|--------|-------|-------|-------|-------|-------|
| Cu    | 1.5    | 1.0    | 2.25   | 0     | 0.15  | 0.25  | 0.015 | 0.1   | 0.05  |
| Al    | 1.5    | 1.0    | 2.25   | 0     | 0.15  | 0.25  | 0.015 | 0.1   | 0.05  |
Then for comparison, the tensile behavior of 0.35 mm thick Cu/Al strip with larger thickness of IMCs layer to be 20 \( \mu \text{m} \) is also molded. The predicted evolution of equivalent stress is given in figure 9. It can be seen that similar failure process occurs, while degree of interface concentration in IMCs layer increases and causes earlier fracture of IMCs layer compared with figure 8, which will decrease the ductility of the clad strip.

3.2. Quantitative prediction of tensile behavior variation with IMCs thickness for 0.35 mm thick clad
For quantitative estimation of the effect of IMCs thickness on the tensile behavior of clad strip, a serials virtue tensile tests of 0.35 mm Cu A T1-1 clad strip with IMCs thickness varied from 2 \( \mu \text{m} \) to 40 \( \mu \text{m} \) were carried out using the proposed model.

The predicted engineering stress-engineering strain curves are shown in figure 10. It can be seen from the figure that, as the thickness of the IMCs layer increases, the elongation of the clad decreases gradually. Especially when the thickness of the IMCs layer is greater than 4\( \mu \text{m} \), the elongation decreases obviously. While, the tensile strength seems changed a little with the variation the IMCs thickness.
3.3. Quantitative prediction of tensile behavior variation with strip thickness with constant 0.79 μm thick IMC layer

For better understanding of IMCs thickness influence on the mechanical property of thin clad strip, the Cu/Al strips with different strip thickness from 0.5 mm to 0.12 mm but same thickness of IMCs layer were also virtually tension tested using the proposed model. The corresponding results for engineering stress-strain curves are shown in figure 11. It can be seen that the predicted elongation decreases gradually with the decrease of the strip thickness, while the simulated tensile strength keeps to a constant and then decreases with further decrease of the strip thickness from 0.2 mm to 0.12 mm.

4. Discussion

4.1. Validation of the modeling

From figure 11, it clearly seen that both ductility and tensile strength has a dramatic drop when strip thickness decreased from 0.2 mm to 0.12 mm compared with that from 0.5 mm to 0.2 mm. And it indicates that the tensile properties could be more sensitive to IMCs layer thickness for ultra-thin clad strips. This predicted results have good agreement with previous experimental finds that bad performance of Cu/Al strip with thickness lower than 0.2 mm by frequent fracture failure under tension loading [15] which validated the proposed model.
4.2. Proposal of a new dimensionless parameter for controlling tensile property of clad strips

The results in figures 10 and 11 quantitatively confirmed the role of IMCs layer thickness in control of tensile mechanical properties of Cu/Al clad strips. For better characterization of such influence, as inspired by the dimensionless parameter $d/t$ (where $d$ is the average grain size and $t$ is the foil thickness) which was reported to control the mechanical property of single metal foil [28], a new dimensionless parameter $w/t$ (where $w$ is the IMCs layer thickness and $t$ is the strip thickness) is tried to proposed in this study to make a controlling relation between the tensile property of the clad metal strip and the thickness of IMCs layer.

Figure 12 gives the predicted tensile strength and elongation obtained from figures 10 and 11 with a relation of newly proposed dimensionless parameter $w/t$. It can be clearly found that the tensile strength and elongation of clad Cu/Al strip decreases with increase of $w/t$, especially when $w/t$ exceeds 0.025, the tensile strength has a dramatical decrease. The parameter $w/t$, which has a physical meaning that the ratio of IMCs layer to the whole strip thickness could be used to reflect the mechanical property of Cu/Al clad strip in this study. The mechanical property of Cu/Al strip can be deteriorated when $w/t$ exceeds certain level (due to thick IMCs layer or ultra-thin strip). This can reasonably answer the experimental finding that bad performance of Cu/Al strip with thickness lower than 0.2 mm under tension loading [15].

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**Figure 10.** The simulated tensile engineering stress-engineering strain curves of 0.35 mm thick Cu/Al clad strips with different thicknesses of IMCs layer.

**Figure 11.** The simulated tensile engineering stress-engineering strain curves of Cu/Al clad strips with different strip thicknesses and same IMCs layer thickness of 7.93 μm.
4.3. A potential application of the proposed method

As mentioned above, the mechanical property can be deteriorated when clad strip thickness decreased bellow 0.2 mm caused by a relative larger \( w/t \) value. Therefore, the tensile property improvement of 0.12 mm thick clad strip is presented here as an example of the potential application of the present study.

Normally the prepared 0.12 mm thick clad strip using process shown in figure 1 was annealed under 693 K for 2 min holding. And the tensile strength and elongation were less than 150 MPa and less than 10% respectively.

According to figure 12, for getting better tensile property, the \( w/t \) value should be less than 0.025. So, for the 0.12 mm thick strip, the total IMC layer thickness \( w \) should be controlled less than 0.003 mm.

![Figure 12. Variation of tensile property of Cu/Al strip with dimensionless parameter \( w/t \).](image)

![Figure 13. The IMC layer thickness evolution in micron with annealing temperature and time (a) Al\(_4\)Cu\(_9\) layer thickness, (b) Al\(_2\)Cu layer thickness, (c) AlCu layer thickness, (d) total IMC thickness.](image)
The evolution of IMC layer thickness with different annealing temperature and time can be obtained using previous proposed IMC growth kinetics model [9], which is shown in figure 13. According to the above analysis, the annealing process should be controlled in the red region. So, a new annealing process under 823 K for 3.5 min holding was carried out. And the tensile strength and elongation of 0.12 mm thick clad trip changed to over 170 MPa and over 14% respectively, which shows great improvement compared with previous results.

Therefore, the established microstructure-based modeling gives a new way for quantitative understanding the factors affecting mechanical property of clad strips. And the new dimensionless parameter $w/t$ offers an efficient way to control the mechanical property of clad strips. As obtained in above paragraph, critical $w/t$ value can be achieved to attain acceptable mechanical property of Cu/Al clad strip by optimize the annealing process.

5. Summary

A microstructure-based model considering both damage evolution of matrix and interface IMCs is proposed for quantitative estimation of tensile mechanical property of Cu/Al clad strips. A new dimensionless parameter $w/t$ is also proposed to be a control index for tensile mechanical property of Cu/Al clad strips. And a critical value of $w/t$ which is 0.025 in this study could exist, exceeding which it is thought to cause deterioration of Cu/Al strip mechanical property. The proposed model and parameter in this study could be used for annealing process optimization and mechanical property control of Cu/Al clad strips with different thickness.

Acknowledgments

This work was supported by the National Natural Science Foundation of China under grant No.51104141 and No. 50971039, and the Fundamental Research Funds for the Central Universities under grant No. N170704015

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