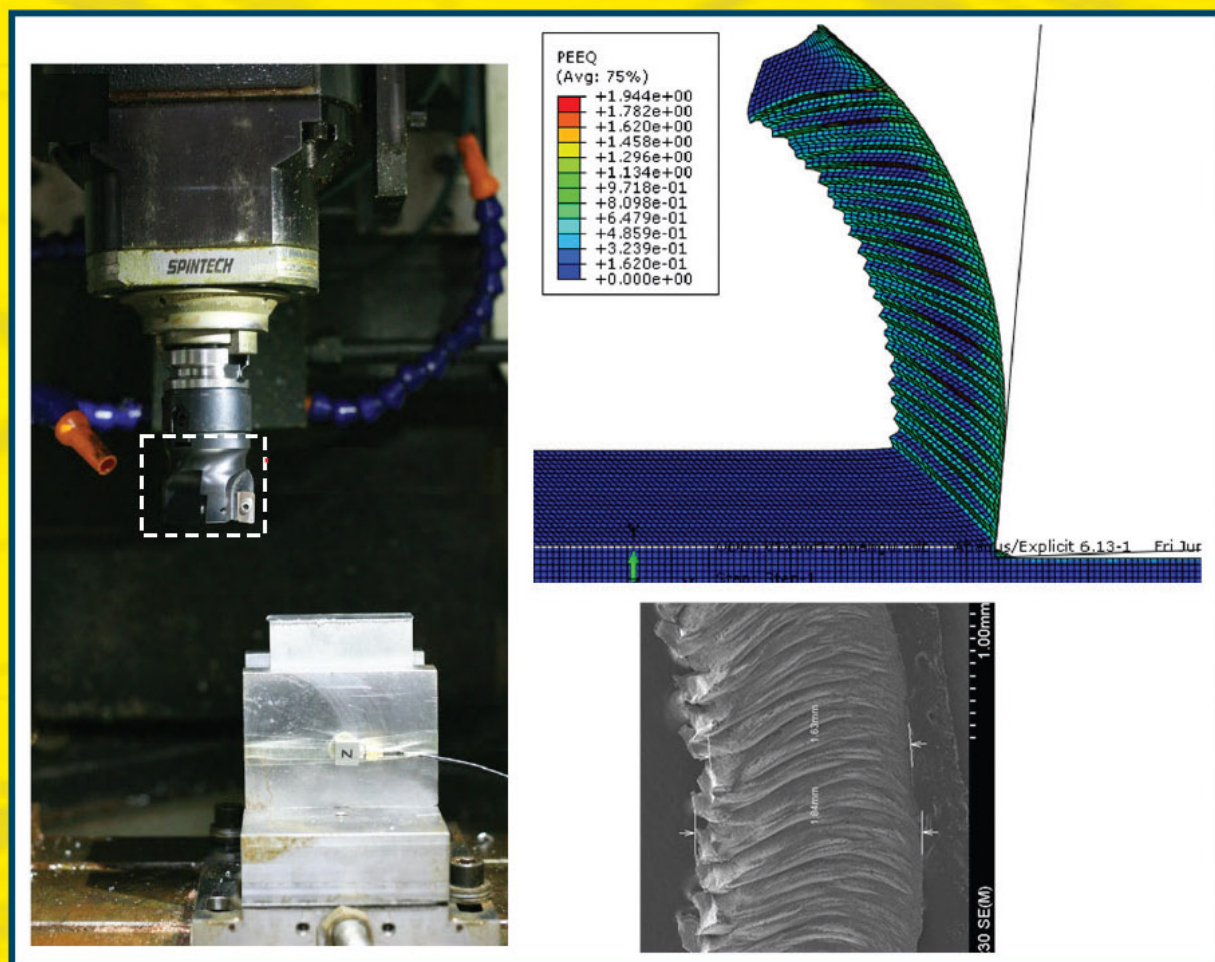


Material, Machines and Methods for Sustainable Development

International Conference on Material,
Machines and Methods for Sustainable
Development (MMMS 2018)



Edited by

Banh Tien Long, Hyung Sun Kim, Kozo Ishizaki,
Nguyen Duc Toan and Nguyen Thi Hong Minh



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Edited by

**Banh Tien Long, Hyung Sun Kim, Kozo Ishizaki,
Nguyen Duc Toan and Nguyen Thi Hong Minh**

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Preface

On behalf of the Organizing Committee, we are delighted to issue the journal which contains papers presented at the first International Conference on Material, Machines and Methods for Sustainable Development (MMMS) from 18-19 May, 2018 in Danang, Vietnam. The MMMS2018 continues enabling researchers from all over the world to share their research and accomplishments as well as review the research currently being conducted by their peers.

The purpose of the journal is to explore and ensure an understanding of the critical aspects which contributed to sustainable development, especially materials, machines and methods. From such understanding, the conference aims at an overall approach to assist policy makers, industries and researchers at various levels to position the local technological development towards sustainable development in the global context, to assist decision making towards a greener approach especially for material, machines and methods.

We would like to express our gratitude to the members of International Organizing and Academic Committees of the Conference, for their hard work and advices which were helpful in maintaining the high level of the Conference. We also thank the Trans Tech Publications Inc. for the publishing the selected papers in the journal of Applied Mechanics and Materials. We hope that all the papers will become a source of valuable information in the scientific work for academics, researchers, engineers and students.

Best regards

Banh Tien Long,

Hyungsun Kim,

Kozo Ishizaki,

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CHAPTER 1:

Advanced Materials, Processing Technologies, Chemical Engineering

Effects of Starting Materials on Preparation and Properties of Pure SiC Ceramics via the HTPVT Method

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Keywords: Silicon carbide; Starting materials; Grain size; Mechanical properties.

Abstract: The method of high temperature physical vapor transport (HTPVT) is an available approach to prepare silicon carbide (SiC) ceramics with high density and high purity. In the present work, α -SiC (6H-SiC) and β -SiC (3C-SiC) powders were used as starting materials respectively to fabricate SiC ceramics with HTPVT process, and the effects of starting materials on nucleation, density, microstructure and mechanical properties of SiC ceramics were investigated. It showed that at high temperature, the decomposition rate of β -SiC was higher than that of α -SiC, and at the initial nucleation stage, the average grain size of SiC crystal obtained with β -SiC starting materials was smaller than that with α -SiC starting materials, because higher vapour pressure of gas phase which decomposed by β -SiC starting materials facilitated nucleation and growth of SiC grains. Density of the resulted SiC ceramics using α -SiC and β -SiC as starting materials was $3.16 \text{ g}\cdot\text{cm}^{-3}$ and $3.17 \text{ g}\cdot\text{cm}^{-3}$, indicating close values, while, using β -SiC as the starting materials, the grain size was smaller, consequently, the flexure strength was higher. Increasing growth temperature from 2200°C to 2300°C , the densities and the flexure strength of the SiC ceramics using either α -SiC or β -SiC were decreased.

Introduction

Silicon carbide materials have always played an important role in the high temperature fields, and extreme environments, due to its high strength, outstanding resistance to oxidation, and excellent thermophysical properties [1-3]. However, it is difficult to achieve the densification of silicon carbide without sintering aids due to the low self-diffusion coefficients and covalent bond of Si-C [4]. For the densification of the SiC ceramics, sintering additives are usually necessary, and many sintering methods have been developed such as liquid phase sintering and hot-pressing sintering [5-8], however, the additives remaining in the final ceramics act as a “weak” secondary phase, which decreased the high-temperature strength and irradiation behavior [9, 10]. On the other hand, high purity SiC can be obtained by recrystallization process [11], however, its densification was restrained due to evaporation – condensation mechanism. In order to combine the density and purity of the silicon carbide ceramics, fabrication processing of highly dense pure SiC ceramics via high temperature physical vapor transport method has been developed, based on preparation process of silicon carbide single crystal [12, 13], however, the grain size of these specimens was as large as 1 mm due to difficult in spontaneous nucleation and thus the flexural strength was low.

In the previous researches, α -SiC carbon plate were usually used as starting materials and the growth substrate respectively, and high dense SiC ceramics with a thickness of 8 mm and a bulk density of $3.208 \text{ g}\cdot\text{cm}^{-3}$ without additives had been successfully prepared by the HTPVT method after 4 h of growth. Grain sizes and grain orientations of ceramics were exhibited to change with growth time, and polycrystalline SiC ceramics presented a preferred orientation along the (0 0 0 *l*) planes. The degree of grain size and orientation of the silicon carbide increased with increased growth time, and the growth rate increased with the growth temperature and temperature gradient. Silicon carbide

ceramics exhibited a bending strength of only 290 ± 34 MPa, primarily due to the large grain size, and a Vickers hardness of 29.7 ± 0.6 GPa [14].

In the field of industry, both α -SiC and β -SiC had wide application, however β -SiC could be fabricated at relative lower temperature comparing with α -SiC [15], this had a positive significance for energy conservation, so that it was necessary to select more appropriate starting materials in the fabrication of pure SiC ceramics via the HTPVT method. In this work, α -SiC (6H-SiC) and β -SiC (3C-SiC) were used as starting materials respectively, and graphite paper was used as growth substrate for easy separation of SiC ceramics obtained. Effects of starting materials on nucleation, density, microstructure and mechanical properties of pure SiC ceramics via the HTPVT method were investigated.

Experimental

Materials and methods: The available commercially α -SiC and β -SiC powders (mean particle size 178 μm ; Zaozhuang Liyuan SiC Co., Ltd., Zaozhuang, China) were selected as the starting materials and placed loosely in a graphite crucible with an inner diameter of 50 mm. The obtained SiC ceramics were grown on graphite paper substrate in a medium-frequency vacuum induction furnace (3.5 kHz, Model ZGRS-160/2.55 Jinzhou Electric Furnace Co., Ltd., Jinzhou, China). Radio-frequency induction coil was supplied power to generate heat in the graphite crucible and the starting materials were then heated by the crucible. The temperature at the middle of the crucible was controlled to 2200°C and 2300°C, and the temperature difference between the starting materials and graphite paper substrate was about 200°C. Soaking time was 4 h and argon gas pressure of 4×10^3 Pa in the growth chamber was controlled by a mass flow meter.

Materials characterization: Phase formation of samples was identified by X-ray diffraction (XRD; X'Pert PRO, PANalytical, Netherlands) using Cu K α radiation on the polished sample. The XRD patterns were recorded in the 2θ range of 20–80°, with a step size of 0.01° and a scan speed of 10° min⁻¹. Microstructure of samples was observed using a scanning electron microscopy system (S4800). Average grain size was estimated through the Image-Pro Plus quantitative image analysis software (Version 7.0, Media Cybernetics, USA). The densities of specimens were determined by the Archimedes method. Flexural strengths of specimens were tested via the three-point bending test (Model WDT-10, Tianshui, China) with a support distance of 16 mm and a crosshead speed of 0.5 mm min⁻¹. Vickers hardness was tested with a load of 9.8 N using the OmniMet88-7000 Fully Automated Microindentation Hardness Testing System.

Results and Discussion

Effects of starting materials on decomposition rate

At high temperature, the decomposition and condensation of SiC could be occurred, and the weight loss of starting materials could be obtained by measuring the weight of the starting and remaining source of SiC powders. Fig. 1 shows weight loss of starting materials in different time of 10 min, 30 min, 60 min, 180 min and 240 min at temperature of 2200°C. During the process of evaporation – condensation, the weight loss of β -SiC was always higher than that of α -SiC, demonstrating a higher decomposition rate for the β -SiC, which was beneficent to obtain higher vapor pressure of gas phase such as Si₂C and SiC₂, and to facilitate nucleation and growth of SiC grains. The main reason for this phenomenon was that β -SiC as low-temperature phase had a lower decomposition temperature than α -SiC, β -SiC powders started to decompose at about 1800°C and α -SiC powders started to decompose at about 2100°C, therefore, when temperature increased to 2200°C, the weight loss of β -SiC starting materials was higher, especially in the early stages of 10 min, the difference of weight loss was large. However, there was a temperature gradient about 10–20°C cm⁻¹ in the graphite crucible, after SiC starting materials at the bottom of the crucible decomposed completely, the rest of SiC starting materials could not decompose sufficiently, so that the decomposition rate decreased and the weight loss of starting materials was approached gradually. As

soaking time prolonged, the rest of starting materials in the crucible were recrystallized and reached densification gradually, the process of vapor transport in the growth chamber was restricted, consequently, the pressure of gas phase got decreasing, and finally the growth of SiC ceramics stopped in 4 h.

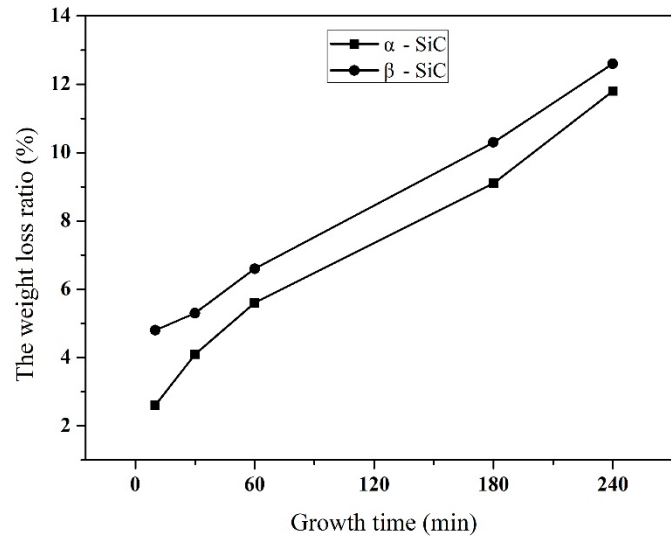


Fig. 1. The weight loss of starting materials with different growth time at 2200°C.

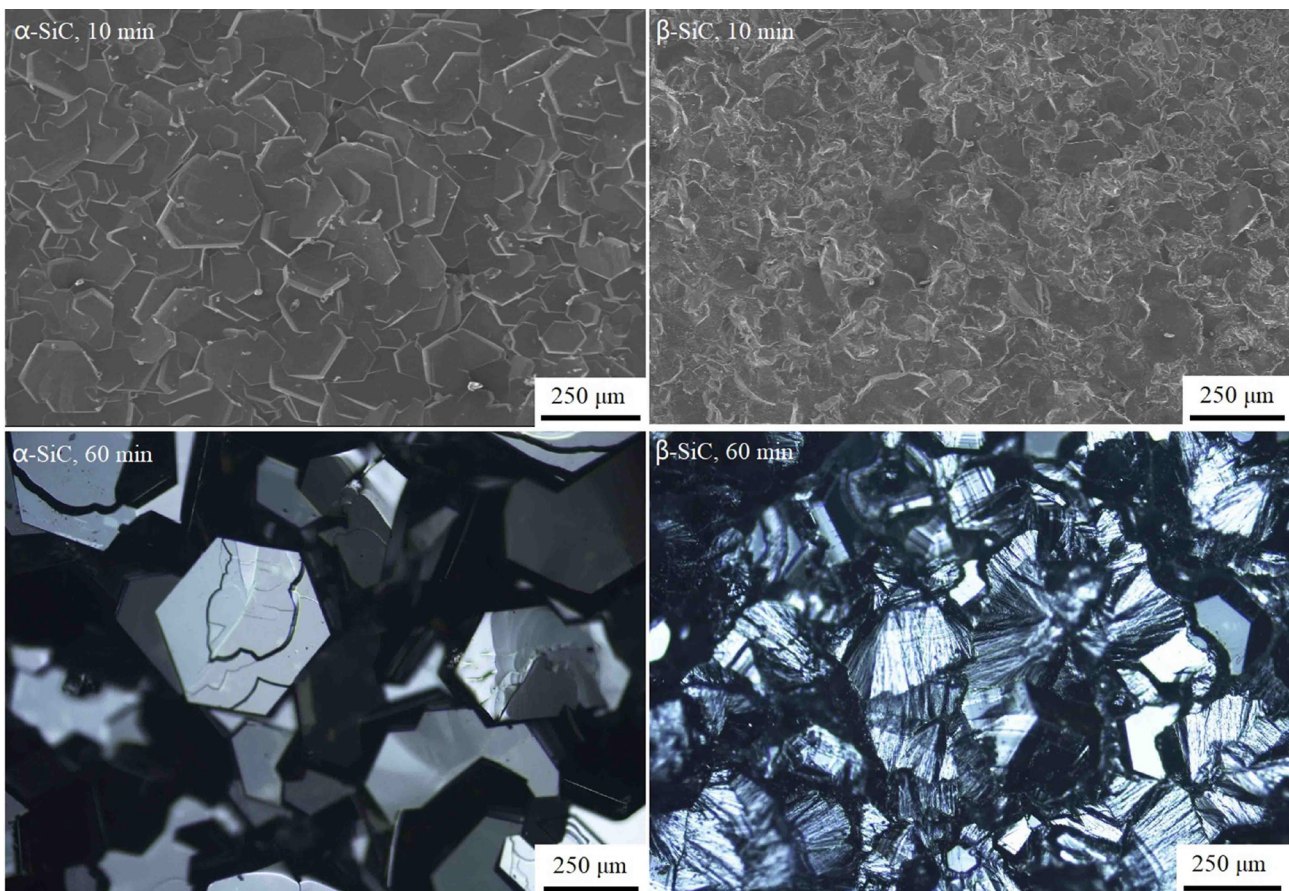


Fig. 2. The morphologies of the as-grown surfaces of SiC ceramics using α -SiC and β -SiC as starting materials with growth time of 10 min and 60 min at 2200°C.

Effects of starting materials on microstructure and phase composition

Fig. 2 shows the morphologies of the as-grown surfaces of SiC ceramics using α -SiC and β -SiC as starting materials with growth time of 10 min and 60 min at 2200°C. After statistics, the average grain size of obtained SiC crystal in 10 min using β -SiC as starting materials was 69 μm , which was smaller than 161 μm for the SiC crystal using α -SiC as starting materials. With increasing the growth time to 60 min, the average grain size was increased, however, the SiC obtained using β -SiC as starting materials still had a smaller grain size. It indicated that in the early stages of growth, β -SiC occupied a higher decomposition rate and the pressure of gas phase in the growth chamber was enhanced, which further facilitated nucleation of SiC grains, resulted in the refined grain size of SiC. The morphology of the as-grown surfaces of SiC grains obtained using α -SiC as starting materials exhibited regular hexagonal, which possibly presented preferred orientation along the (0 0 0 *l*) crystal planes for the SiC grains. The morphology of SiC grains obtained using β -SiC as starting materials exhibited relative irregular, which demonstrated a relative disordered orientation for the SiC grains, as also confirmed by XRD as shown in the Fig. 4.

Fig. 3 presents the morphologies of the as-grown surfaces of SiC ceramics grown for 4 h at 2200°C and 2300°C. With the soaking time of 4 h, the average grain sizes of SiC ceramics using α -SiC and β -SiC as starting materials increased to 501 μm and 307 μm respectively. Increasing growth temperature from 2200°C to 2300°C, change of the grain size of SiC ceramics obtained was limited, while the defects of SiC ceramics increased obviously. Defect density of SiC ceramics obtained using β -SiC as starting materials was high, and the cavity defects with 20~30 μm in diameter on the surface of SiC ceramics maybe lead decrease of the density and mechanical property of SiC ceramics. The reason of defects increasing was that SiC ceramics obtained could decompose again at higher temperature, especially 3C-SiC existed in SiC ceramics obtained using β -SiC as starting material was easily decomposed and resulted in more defects due to the low-temperature phase.

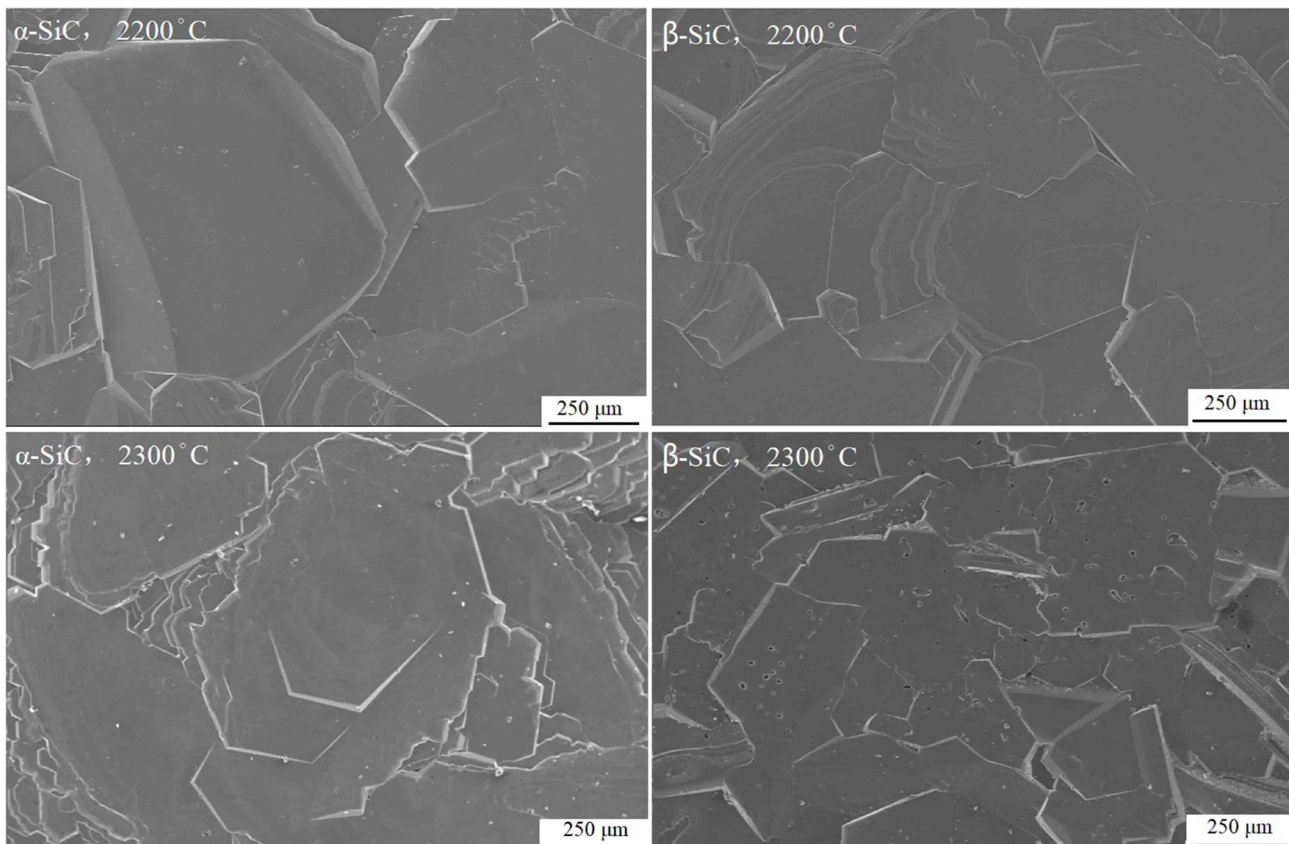


Fig. 3. The morphologies of the as-grown surfaces of SiC ceramics using α -SiC and β -SiC as starting materials grown for 4 h at 2200°C and 2300°C.

Fig. 4 shows the phase composition of SiC ceramics obtained at 2300°C for 4 h with α -SiC and β -SiC starting materials. By the XRD analysis, it indicated that SiC ceramics obtained with α -SiC starting materials were identified as pure 6H-SiC, and mainly showed preferred orientation along the (0 0 0 6) and (0 0 0 12) crystal planes; however, SiC ceramics obtained with β -SiC starting materials were identified as mixture of 6H-SiC and 3C-SiC phase. As low-temperature phase of 3C-SiC, it was partly decomposed into silicon vapor and graphite at higher temperature, so that the XRD analysis showed the existence of graphite.

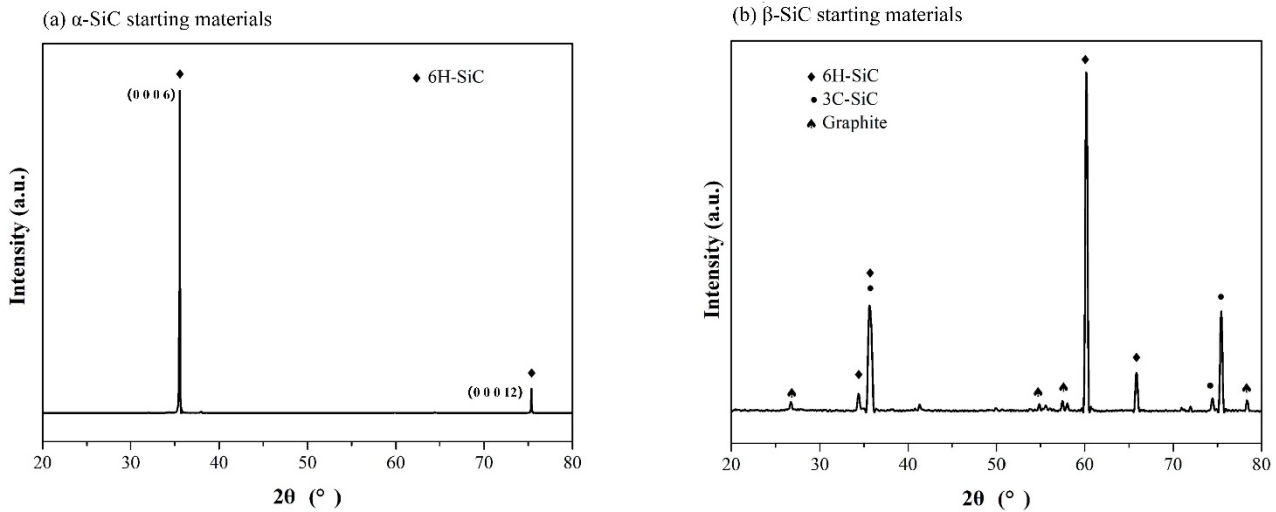


Fig. 4. XRD patterns of SiC ceramics obtained at 2300°C with α -SiC and β -SiC starting materials.

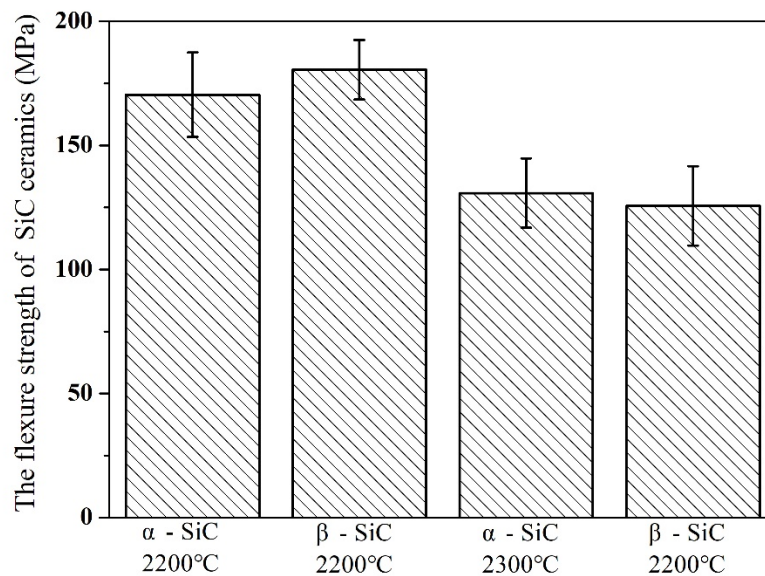


Fig. 5. The flexure strength of SiC ceramics obtained at 2200°C and 2300°C for 4 h with α -SiC and β -SiC starting materials.

Effects of starting materials on density and mechanical property

Fig. 5 presented the mechanical property of SiC ceramics obtained at 2200°C and 2300°C for 4 h with α -SiC and β -SiC starting materials. As shown in the Fig. 5, the flexure strength of SiC ceramics using β -SiC as starting materials obtained at 2200°C was about 180.5 ± 19.7 MPa, which was higher than 170.4 ± 14.3 MPa of SiC ceramics using α -SiC as starting materials. This was due to the smaller grain size of the SiC ceramics obtained from β -SiC, which was in consistent with that the flexure strength of ceramic materials depended largely on the grain size and decreased with increasing grain

size. On the other hand, the flexure strength of SiC ceramics obtained by β -SiC at temperature of 2300°C was lower than that by α -SiC, which was mainly due to the existence of defects in SiC ceramics through partial decomposition of 3C-SiC at higher temperature.

The densities of the SiC ceramics are listed in Table 1. At 2200°C, the samples prepared by HTPVT method all exhibited higher density of more than 98%. With increasing the soaking temperature from 2200°C to 2300°C, the densities of SiC ceramics obtained by α -SiC and β -SiC starting materials decreased to 3.14 g·cm⁻³ and 3.04 g·cm⁻³, respectively, which was attributed to the existence of defects with 20 - 30 μ m in diameter caused by decomposition of 3C-SiC at higher temperature in the SiC ceramics obtained.

Table 1. The densities of SiC ceramics.

Growth temperature (°C)	Starting materials	Density (g·cm ⁻³)	Relative density (%)
2200	α -SiC	3.17	98.7
	β -SiC	3.16	98.4
2300	α -SiC	3.14	97.8
	β -SiC	3.04	94.7

Conclusion

SiC ceramics obtained with α -SiC and β -SiC as starting materials have been successfully prepared by the HTPVT method. During the whole process of evaporation – condensation, the weight loss of β -SiC was higher than that of α -SiC, which indicated a higher decomposition rate for the β -SiC starting materials. The high vapor pressure of SiC gas phase was facilitated nucleation and growth of SiC grains, contributed to refinement of the grain size and enhancement of the flexure strength of SiC ceramics. SiC ceramics obtained by α -SiC were identified as pure 6H-SiC with preferred orientation along the (0 0 0 6) and (0 0 0 12) crystal planes; that by the β -SiC were identified as mixed 6H-SiC and 3C-SiC. Increasing soaking temperature from 2200°C to 2300°C, the densities and the flexure strength of SiC ceramics using α -SiC and β -SiC as starting materials were almost decreased, particularly the degree of reduction of density and flexure strength of SiC ceramics using β -SiC as starting material was even greater.

Acknowledgements

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Effect of Residual Stress and Microcracks in Chrome Plating Layer to Fatigue Strength of Axle-Shaped Machine Parts

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Keywords: Fatigue strength, residual stress, chrome plating, X-ray diffraction and microcrack.

Abstract. Chromium plating is used widely in industry to enhance wear, abrasion resistance and to restore the dimensions of undersized parts. However, tensile residual stress always exists in chrome layer because of hydrogen embrittlement so it affect to mechanical properties of the chromium plating machine element, especially in fatigue strength. In this paper, effect of residual stress together with microcracks in chrome plating layer to the fatigue strength was studied. The sample (AISI 1045 steel) was plated with 10 and 60 micrometers thicknesses and residual stress in chrome plating layer was determined by X-ray diffraction technique (Cu-K α radiation). The results showed that as chromium layer thicknesses increased, tensile residual stress decreased and microcrack density increased. Consequently, the fatigue strength went down when chromium layer thicknesses increased.

Introduction

It is recognised that hard chromium plating can increase surface hardness, resistance to wear and corrosion and low coefficient of friction, and it has been extensively used to improve properties of parts such as shafts, piston rings, engine valve stems, aircraft landing gear and engine parts, etc. In particular, they have been widely used for restoring worn, undersized parts. The main objective of hard chrome plating is to increase the service life of the coated parts [1]. The good corrosion and wear resistances due to passivation phenomenon, by forming a chromium oxide film Cr₂O₃, which is a barrier to hydrogen, preventing hydrogen embrittlement.

Another important properties of hard chrome plating layers are existence of tensile residual stress and microcracks. Many researches has investigated the effect of chromium plating on fatigue life of high strength steel. The tensile residual stress in chrome plating layers are relieved by local microcrack during electroplating [2]. According to Hammond and Williams, the magnitude of tensile stress varies depending upon the plating solution and operating conditions [3]. These residual stresses can be partially relieved by cracking and, therefore, the density of the cracks pattern present in the coating will be intimately associated with the decrease in fatigue properties. The fatigue strength decrease when hard chrome plated on AISI 4340 and the tensile residual stress was shown [4]. The fatigue strength improvement of a hard chromium plated AISI 4140 steel using a plasma nitriding pre-treatment also was study. It was found that hard chromium-plated specimens with a coating layer of 23 ± 2 μ m thickness showed approximately 33% reduction in fatigue strength when compared to quenched and tempered specimens. An application of the plasma nitriding pre-treatment before the plating process was effective in improving the fatigue performance of HC-coated steel [5].

The objective of this study was to investigate the influence of residual stress and microcracks in chrome plating layer to the fatigue strength of axle-shaped machine parts which was made by medium carbon steel AISI 1045. The thickness of the coating layer is an important parameter, depending on the current density, temperature and time of plating. The 1045 samples are plated with various thickness to have many stress level and crack densities. Normally, thickness from 10 to 60 micrometers is required for abrasion resistance [6]. The microstructure of the layer are also observed by SEM. The residual stress are determined using the X-ray diffraction method.

Theoretical Background

Fatigue. The relationship between stress and cycles (S-N curve), as shown in Fig. 1, is used to investigate the influence of plating thickness to the fatigue strength in this research.

$$S = f(N) \quad (1)$$

The Basquin's equation is used to determine fatigue strength as [7]:

$$\sigma_a = \sigma'_f (2N_f)^b \quad (2)$$

where σ_a is stress amplitude; σ'_f is fatigue strength coefficient; $2N_f$ is the number of reversals to failure; and b is Basquin's exponent ($b = -0.05 \sim -0.12$).

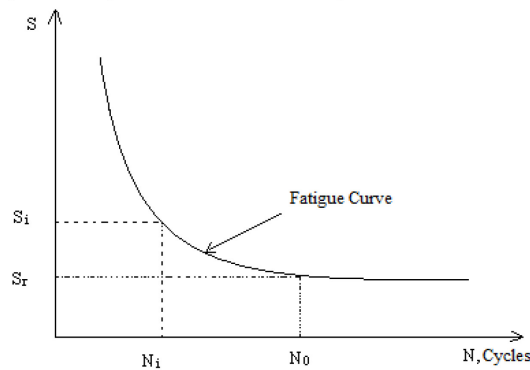


Figure 1. Wöhler Curve [7]

Determination of residual stress using X-ray diffraction. The most common and accurate technique for residual stresses determination is X-ray diffraction (XRD). For elastic materials, lattice strain, representing the stress, can be determined through the shift of peak position 2θ of the X-ray diffraction lines (Fig. 2). The relation between lattice spacing and the peak position are shown in the Bragg law equation as [8]:

$$\lambda = 2d_{hkl} \sin \theta \quad (3)$$

where λ is the X-ray wave length, d_{hkl} is the lattice spacing of the (hkl) plane, and θ is the Bragg's angle.

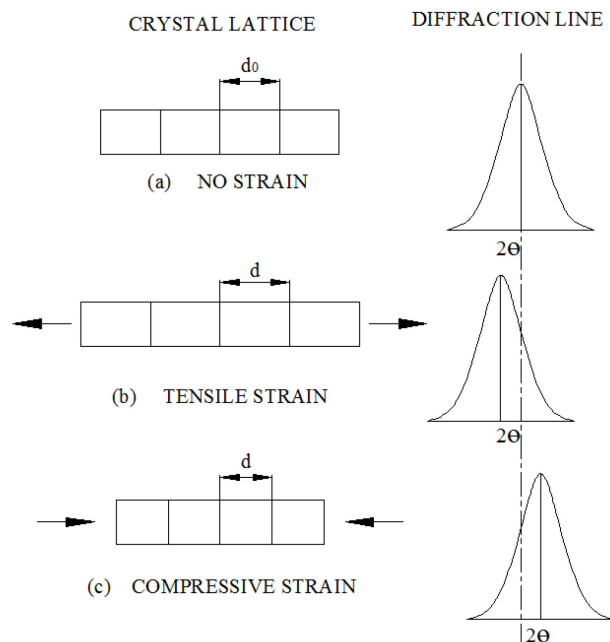


Figure 2. Diffraction line expanded due to stress

The peak position shifts to the left for a tensile stress and to the right for a compress stress. Now, the lattice spacing d_{hkl} is compared with that for an unstressed lattice spacing d_0 to determine the lattice strains. Therefore, the strain and stress is determined from the Bragg angle θ . The Hooke's law representing the relation between strain and stresses is as bellow [8]:

$$d_{hkl} = d_0 \frac{1+\nu}{E} \sigma_\phi \sin^2 \psi - d_0 \frac{\nu}{E} (\sigma_x + \sigma_y) + d_0 \quad (4)$$

where d_{hkl} and d_0 are stressed and unstressed lattice spacing; E , ν and σ_ϕ are the Young's modulus, the Poisson's ratio, for chrome, $\nu = 0.23$ and $E = 271 \text{ GPa}$; σ_x , σ_y , and σ_ϕ are the residual stresses in the x -, y - and ϕ -directions, respectively. Equation (4) demonstrates a linear variation of the plane spacing versus $\sin^2 \psi$ with the slope of the m :

$$m = d_0 \frac{1+\nu}{E} \sigma_\phi \quad (5)$$

Therefore, the residual stress can be determined as:

$$\sigma_\phi = \frac{m}{d_0} \left(\frac{E}{1+\nu} \right) \quad (6)$$

Residual stress of chrome plating layer was determined by X'Pert Pro Panalytical system in which the X-ray source is Cu-K α radiation. {211} diffraction plane with the diffraction angle 2θ from 82° to 83° were used. Three tilts of $\sin^2 \psi = 0, 0.05$ and 0.1 in the $\sin^2 \psi$ diagram were used for to compute the residual stress.

Experiment and Results

Experiment

Experimental equipment. Bending fatigue experiments were conducted in Research group Environmental and Mechanical Engineering (REME Lab, HCMC University of Technical Education). The principle of rotating bending fatigue are shown in Fig. 3. Bending fatigue experiments were based on ISO 1143:2010 standard and performed on bending fatigue testing in REME Lab. The velocity of fatigue testing is approximately 2,000 rpm.

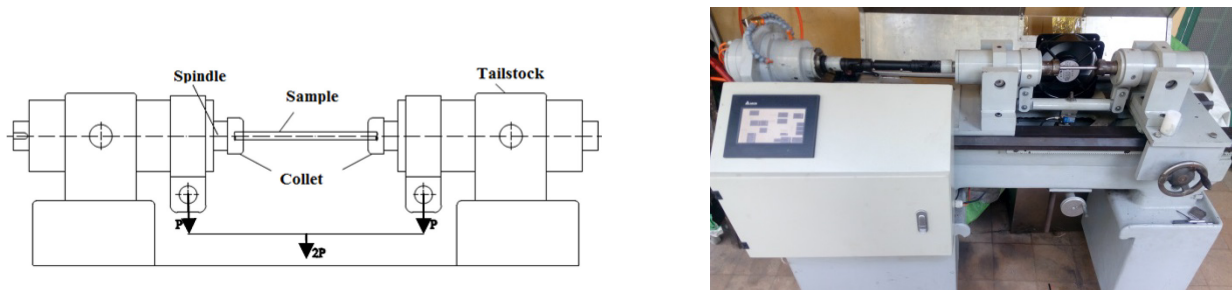


Figure 3. Principle and machine of rotating bending fatigue

Sample experiment. According to the ISO 1143:2010 standard and specifications of the bending fatigue equipment, the samples are manufactured from AISI 1045 steel and its dimension is showed in Fig. 4. Table 1 is the chemical compositions of AISI 1045 sample.

Table 1. Chemical compositions of sample

C	Si	Mn	P	S
0.46%	0.26%	0.61%	0.025%	0.038%

The specimens were quenched at 840°C in water and tempered at 600°C to obtain hardness of 28 - 30 HRC to obtain the heat treatment process of shafts. Average superficial roughness in the reduced section of the specimens was $R_a = 0.77 \mu\text{m}$.

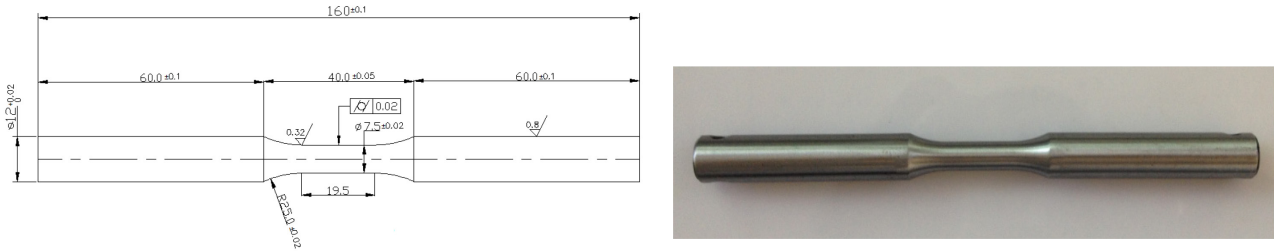


Figure 4. Specimen for fatigue experiment

Table 2 shows mechanical properties of fatigue sample.

Table 2. Mechanical properties of sample

Sample	Yield strength σ_c (MPa)	Tensile strength σ_b (MPa)	Elongation (%)
01	1066.56	1123.89	14.925
02	1090.04	1135.36	14.766
03	1091.42	1168.38	13.433
Average	1082.67	1142.54	14.374

Creating chromium electroplating layer. Plating process is performed in the REME Lab. Experiment equipment are including electroplating power (current maximum 50A), thermocouples for measuring the temperature of the bath and clamping equipment. Plating solution is shown in Table 3 and the plating process is performed in the room temperature of about 32 - 33°C. The thickness was measured by Positector - Defelsko (Eddy current testing). The deviation of this equipment is 1% of the actual thickness. Tables 4 show the average of 5 measurement values for each sample.

Table 3. Plating solution

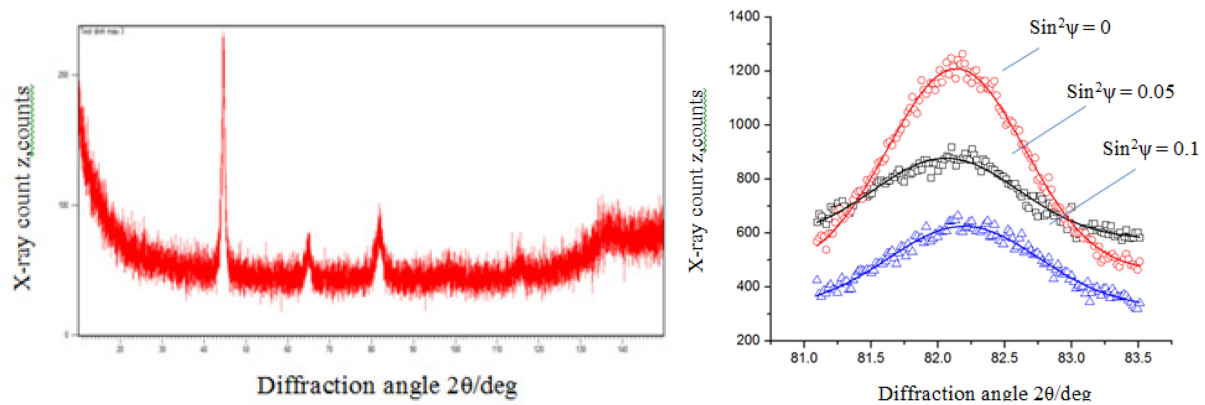
Component of solution	Content(g/l)
CrO ₃	250
H ₂ SO ₄	2.0
Cr ³⁺ Concentration	3 - 6
Temperature, °C	55
i_a (A/dm ²)	30
H (%)	12
CrO ₃ /H ₂ SO ₄ ratio	100/1
S _a /S _c area ratio	1/1 - 2/1
Plating speed, $\mu\text{m/ph}$	0.15 - 0.60

Table 4. Results of thickness plating layer

Specimen	Thickness t (μm)
1	10 ± 1
2	62 ± 3

Results

Residual stress. Figure 5 shows the XRD pattern and the diffraction lines of the chrome plating layer. The lines were fitted by Gaussian curve to determine the peak position 2θ . The stressed lattice spacings were calculated from the peak position using Bragg's law.



a) XRD pattern of chrome plating

b) The diffraction lines fitted by Gaussian functions

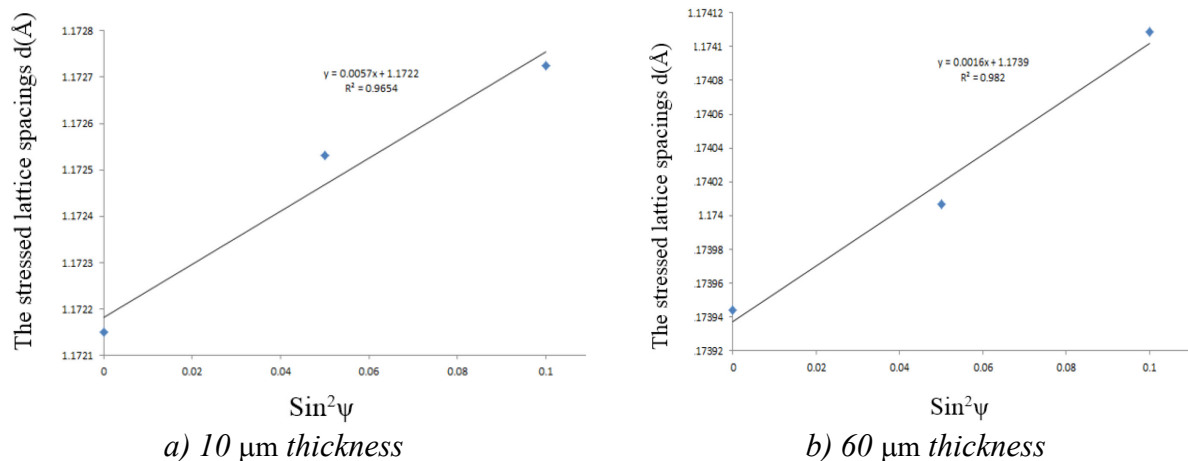
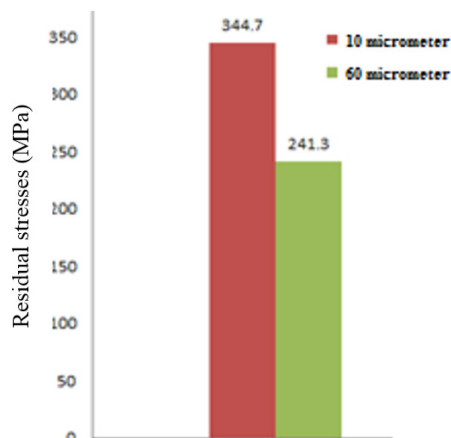
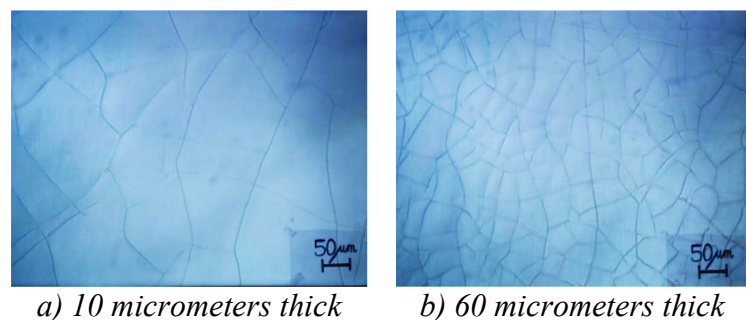
Figure 5. Diffraction lines 10 micrometers thick chrome plating layer**Table 5.** Peak position p (deg.)

Thickness	$\sin^2\psi$		
	0	0.05	0.1
10 μm	82.194	82.136	82.060
60 μm	82.042	82.019	81.944

Table 6. The stressed lattice spacings d ($\text{\AA}$)

Thickness	$\sin^2\psi$		
	0	0.05	0.1
10 μm	1.1721	1.1728	1.1737
60 μm	1.1739	1.1742	1.1750

Fig. 6 shows the relationship of $d\text{-}\sin^2\psi$ for chrome plating with 10 and 60 micrometers thickness.

a) 10 μm thicknessb) 60 μm thickness**Figure 6.** $\sin^2\psi$ diagram**Figure 7.** Residual stresses of Chrome plating layer

a) 10 micrometers thick

b) 60 micrometers thick

Figure 8. Microcrack density of chrome plating layer

It was observed that the tensile residual stresses always exist in chrome plating layers. The stress decrease with the increasing thickness layer [9]. The reason is that during plating, hydrogen atoms always exist in chromium crystalline lattice. At the end of the plating process, the hydrogen atoms

exhaust causing residual stress and microcrack in chrome layer. The microcracks form during electroplating when the tensile exceeds cohesive strength of chrome. Microcrack density increase with increasing thickness [10,11]. There is high tensile stress and low microcrack density in thin layers of 10 μm . However, the exhausting hydrogen and microcrack density are larger than those in the 60 μm thickness. This increase in cracks reduces the stresses in the coating layer, as shown in Fig. 7 and Fig. 8.

Fatigue. Bending fatigue testing was performed for two types of samples: substrate and chrome-plated with thicknesses of 10 and 60 μm . The fatigue curves for the two types of samples are shown in Fig. 10. From the above results, we draw the fatigue curve of samples. The S-N curve for the Basquin model is established for each thickness.

The microscope structure of the chrome-plated samples in Fig. 8 shows that the microcrack exists in chrome plating layer. Fatigue fracture is caused by the cracks from surface where applied the highest stress. This is a hard chrome coating on the outside of shaft where the cracks are present when plating and developing into cracks while operation causes the component machine to break down prematurely.



Figure 9. Chrome coating sample was fractured at $\sigma = 450$ MPa

Table 7. The parameters of Basquin's equation

Sample	σ'_r (MPa)	b
Substrate	2,290	-0.106
10 μm /chrome plated	1,836	-0.094
60 μm /chrome plated	1,229	-0.071

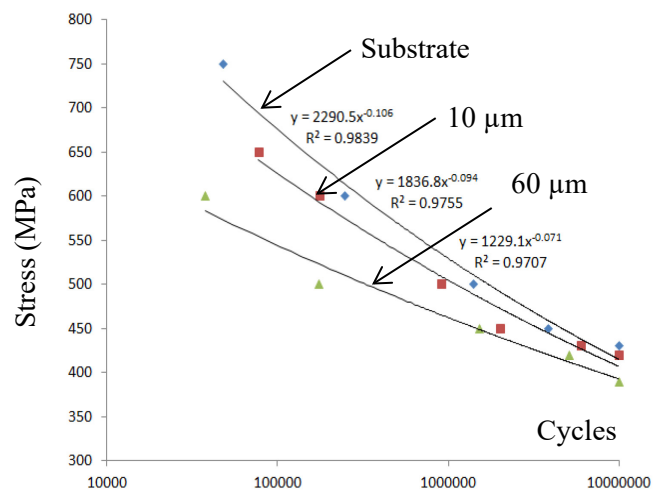


Figure 10. Fatigue curves of samples

The fatigue curves of samples in Fig. 10 show the fatigue life of the chrome coating and uncoated samples. It is possible to observe the great decrease in fatigue strength due to chrome coating by about 87%. Specifically, at stress $\sigma = 450$ MPa, the fatigue life for uncoated sample is 1,398,117 cycles while fatigue strength for 10 μm and 60 μm thickness coated sample are 913,376 and 174,874 cycles, respectively.

Comparison of fatigue limits with the substrate shows that the fatigue limit of chrome-plated sample is lower than those of substrates and decrease with coating thickness. The thickness of the 10 μm coating was $\sigma_{-1} = 420$ MPa (2.3% lower than those of substrate) and the 60 μm coating thickness was $\sigma_{-1} = 390$ MPa (9.3% lower than those of substrate).

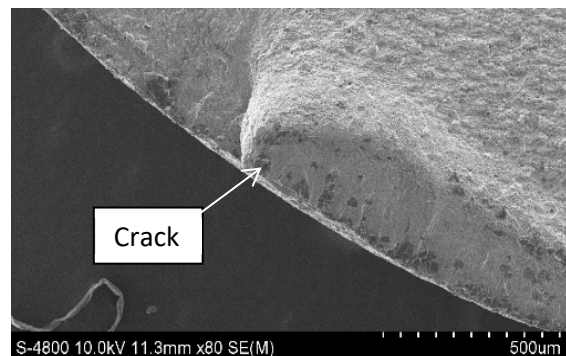
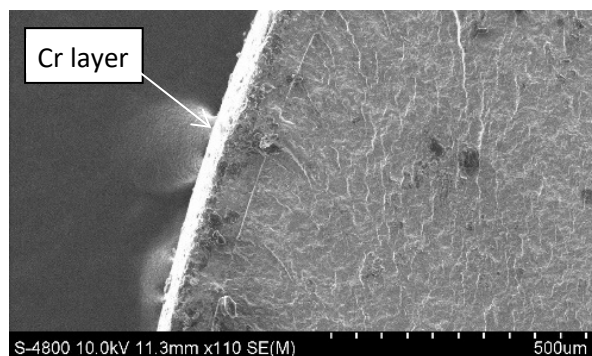


Figure 11. The fracture surface of chrome coating sample with 60 μm chrome plating thickness using the SEM method, $\sigma = 450$ MPa

The SEM results show that fatigue destruction of the specimen originates from the surface of the specimen. For chromium-plated samples, the initial crack that exists in the chrome-plated layer under cyclic loading is the main cause of reduced fatigue strength.

Conclusions

Fatigue function for substrate of AISI 1045 steel and 10 and 60 μm thickness chrome plating were determined using Basquin's model.

The effect of the chrome coating reduce the fatigue strength of the AISI 1045 steel. The reason is hard chrome layer breaks early and leads to premature destruction of the sample.

The coating layer is thicker and the greater microcrack. In the chromium layer there is always exist tensile residual stress and microcrack. For the thicker the plating, the tensile stress decreases but the crack density increases. This is the main cause of fatigue decrease. For the 10 μm chrome plating, the fatigue limit was $\sigma_{-1} = 420 \text{ MPa}$, down 2.3% compared to uncoating, while the 60 μm coating that was $\sigma_{-1} = 390 \text{ MPa}$, down 9.3% compared to uncoated.

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Study on Microstructure and Properties of Cu-9Ni-6Sn Alloy Applied for Electric Measurement

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Abstract. High strength elastic alloy has an important role in the manufacture of electrical equipment and machine building. Along with the development of information and telecommunications technology, the application of this alloy in electrical and electronic equipment is very increasingly. Even on boards or connectors of common electrical equipment such as computers, cell phones, these connectors are usually made of high strength elastic copper alloy. The design required features are small, precision built, with high mechanical strength and elasticity, heat resistance, abrasion and corrosion resistance in the operating environment to ensure its/the power and signal stability for a long time. The design trend is reduced in size but still assure the equipment quality is growing significantly and the corporation to manufacture equipment is researched thoroughly. Accordingly, this article presents the research results about the alloy copper with 9%Ni and 6%Sn which has the high elastic strength and elasticity properties of elastomers after heat treatment. The properties of the microstructure, hardness, conductivity, dry friction coefficient, the corrosion resistance of the alloy from which to determine the parameters for the materials selection procedure and the design of the manufacturing of magnetic contact of this alloy. The results of the study show that after heat treatment and deformation, the Cu-9Ni-6Sn has a strength of alloy up to 1200MPa, the elastic limit is 1100MPa and the conductivity is 8.4%IACS, respectively. The values of this characteristic are consistent with the working conditions of the electrical contact. With the deformation process combined with the heat treatment process, the results of our research group created a single-phase homogeneous microstructure that is chemically stable with the spinodal decomposition in appropriate to the treatment of aging process. By modern methods, this paper demonstrates the durability of the alloy due to the spinodal decomposition during aging treatment at 350°C. This structure of spinodal decomposition is about 20-40nm in size, dispersed throughout the entire cross-section of the sample.

Introduction

High-strength elastic copper alloys play an important role in the manufacture of electrical components and in machine building. Along with the development of information technology, telecommunications technology, the application of this system in electrical and electronic equipment has been increasingly developed on boards or common electrical equipment connectors such as computers, cell phones and the others which are usually made of high strength elastic copper [1, 2]. The design required features are small, precision built, have high mechanical strength and elasticity, heat resistance, abrasion and corrosion resistance in the operating environment to ensure power and signal stability for a long time. The design trend is reduced in size but still ensure the equipment quality is growing significantly and the corporation to manufacture equipment is researched thoroughly. The inventions in this field are applied in the information technologies that are changing our everyday lives [2].

High durability copper alloys are commonly used in electrical equipment such as Cartridge copper (Cu-30Zn), copper phosphate (C51000, Cu-5Sn-0.2P), beryllium copper (C17200; Cu-2Be) Corporal copper (C72700; Cu-9Ni-6Sn, C729000; Cu-15Ni-8Sn). These two systems of copper

alloys are beryllium and spinodal copper which are two copper alloys that are heat-treated and have the highest elongation and elasticity [3-5]. Its hardness, tensile strength and elastic limit are up to 40HRC, 1300-1400Mpa and 1200Mpa, respectively [2]. Beryllium Copper is an alloy of various properties suitable for high-end applications in electrical equipment but with high production costs and demanding manufacturing technology, beryllium is a toxic element that causes harm to human health so it is limited use over the world. In present, the beryllium system is completely imported, so the need of finding a comparable copper system with lower production costs can be the matter of urgency in our country [6].

Identified requirements with an alloy when designing a contact point are as below:

- Good durability and elasticity.
- Low resistance and good conductivity
- Abrasion by external agents
- Small contact friction coefficient

High melting temperature and vaporization

- Its oxide has a large electric charge
- Good machining ability
- Low cost

Some few alloys can adequately satisfy these factors when selected material is identified on the technical requirements for selecting the right material. Some high strength alloys are used to make electrical contacts:

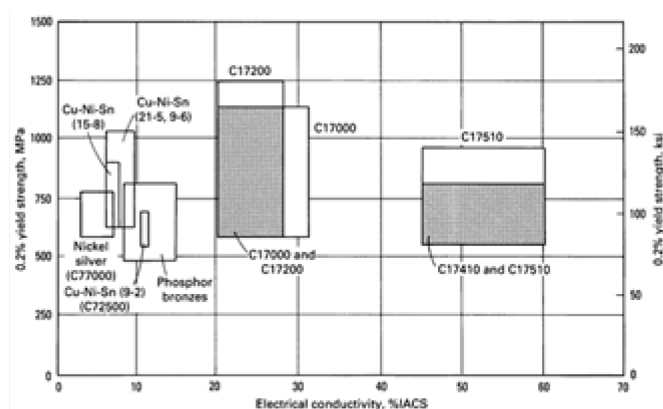


Fig.1 Comparison of conductivity and the elastic limit of some copper alloys for fabrication [1].

The spinodal Cu-Ni-Sn alloys are being invested in recent years to replace beryllium in the manufacture of electrical appliances. In this paper, the authors present the microstructures and properties of Cu-9Ni-6Sn alloy for the manufacture of electrical contact devices.

Experimental Procedure

This alloy was melted in an induction furnace, the composition of alloys are shown in Table 1:

Table 1 The chemical composition of Cu-9Ni-6Sn

Cu	Zn	Pb	Sn	P	Mn	Fe	Ni	Si	Mg	Al	S
83.6	0.0423	0.005	6.35	0.003	0.0025	0.046	9.71	-	-	0.002	0.15

The casting alloys are homogenized, quenched to room temperature, and aging. The samples of Cu-9Ni-6Sn were annealed at 800°C for 2 hours to form a homogenized solid solution, and then aged 350°C for 2 hours. The rate of samples was deformed before aging with cold deformation is 40%.

Additionally, the alloy was characterized for microstructure, hardness before and after treatment at Hanoi University of Science and Technology (HUST).

About the results of Xray, all experimentals were conducted at the Shimane University, Japan.

In addition, alloys have been identified for tensile strength, elastic limit, electrical conductivity at the Military Science and Technology Institute (MSTI).

Results and Discussions

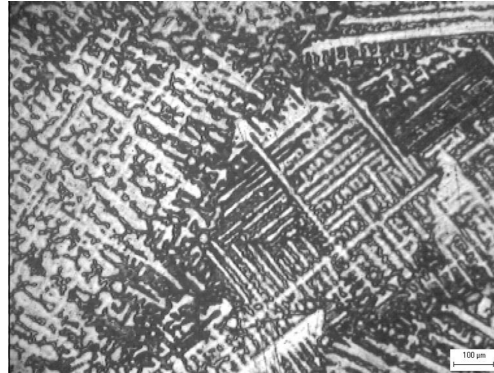


Fig. 2 *Microstructure of Cu-9Ni-6Sn alloy after casting, X50*

Figure 2 shows Cu-9Ni-6Sn microstructure of which the post-casting's microstructure is heavily bombarded by the crystallization of Ni and Sn during freezing due to varying freezing temperatures. The arrangement of large branches and branches are created due to the uneven distribution of Ni and Sn during crystallization.

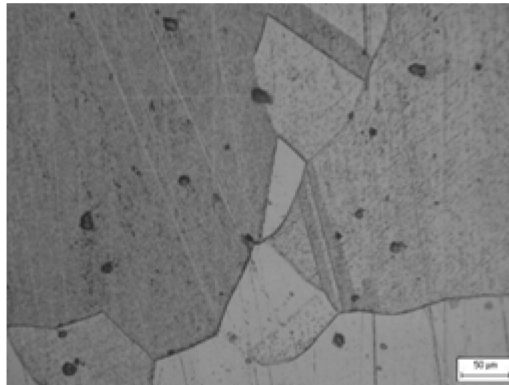


Fig. 3 *Microstructure of Cu-9Ni-6Sn alloy after quenching, X200*

The sample after incubation for 2h at 800⁰C showed that the branches phenomenon was eliminated. The microstructure is essentially phase α but shows an irregular contrast due to the diffusion of the elements in the particle. The concentration of elemental alloys between the particles is not uniform. Diffusion also helps to redirect the particles that are the result of the bundle removal of branches of which microstructure of the base is uniformly.

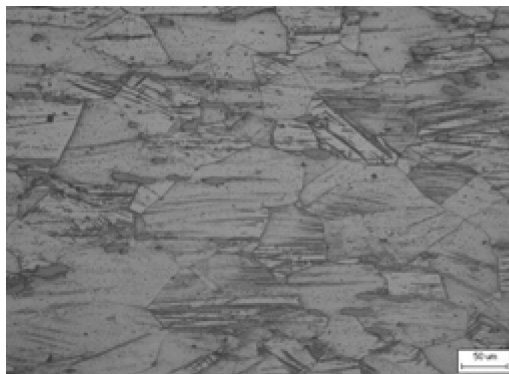


Fig. 4 *Microstructure of Cu-9Ni-6Sn alloy after quenching, hot rolled, then cold rolled with 40% deformation, X200*