The Effect of Aluminum on Microstructure and Mechanical Properties of ATI 718Plus Alloy

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In this reported study, the influence of aluminum concentration on the microstructure and mechanical properties of alloy ATI 718Plus was investigated. The experimental results revealed that the content of aluminum has a significant effect on the solvus temperature and the chemical compositions of the γ’ and δ phases of ATI 718Plus. The experimental data showed that an increased concentration of aluminum in the alloy caused the tensile strength to increase, but had a negative impact on the metal’s ductility. The distribution of aluminum in the matrix and phases of the alloy were systematically examined using the atom probe tomography (APT) technique. The precipitation of phases in the alloy resulting from heat treatment was predicted using the JMatPro 6.0 software application and the predicted results were verified experimentally using electrolytic phase separation followed by microchemical analysis. In addition, a more detailed analysis of the size distribution of the γ’ and crystal structure of the δ phase in the alloy was performed using SEM and XRD analysis. A full discussion of the effects of aluminum content on the mechanism of the generation of the alloy microstructure and evolution of the tensile properties is included in this report.

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1. Introduction

The ATI 718Plus alloy (hereafter referred to as 718Plus) is a newly developed nickel-based superalloy produced by ATI Allvac Company designed to replace Inconel 718 (hereafter referred to as 718) for use at higher temperatures in static and rotating applications in gas turbine engines. The operating temperature capability of the 718Plus is at least 328 K (55°C) higher than the alloy 718. Using the chemical composition of alloy 718 as the base, the Al/Ti and Al+Ti content were modified and the Fe content was decreased from 18% to 10%, together with addition of 1% W and 9% Co to form alloy 718Plus.¹–⁶ It is interesting to note that the strength of the resulting alloy is converted from predominately γ to γ’ after modifying the composition and the operating temperature improves to 977 K (704°C). These improvements are not realized at the expense of the attractive properties of the alloy which still retains the excellent mechanical properties, good fabricability of alloy 718. This highly desirable combination of characteristics places 718Plus in a desirable position to effectively fill the longstanding gap between the two most widely used wrought superalloys; 718 and Waspaloy. Comparisons have been made to both 718, Waspaloy and other superalloys illustrating that alloy 718Plus is the best available candidate to sustain the advances in engine development made possible by the forty-plus year life of alloy 718.⁷

The content of aluminum in superalloy is usually needed to precisely control in metallurgical operation because of its significant effect on the mechanical properties. Aluminum is the principal element forming the γ’-Ni₃(Al,Ti) strengthening phase in most superalloys. But the difference is that the γ’ phase in 718Plus also includes Nb, which is considered to promote the slow precipitation rate of γ’ phase in the alloy, which is why 718Plus has good welding and hot working ability compared to other γ’ strengthening superalloys.⁸ It has been found that the best combination of mechanical properties and thermal stability occurred with a 4% Al+Ti level and a Al/Ti(at%) ratio of about 4 in the 718Plus alloy using modeling calculations employing the JMatPro 2.0 program.⁹ In spite of the development of alloy 718Plus, the mechanism of the effect of aluminum on microstructure and properties of this alloy is still not well understood. In this study, alloy 718Plus has been evaluated to determine the effect of aluminum content adjustments on microstructures and tensile properties of the resulting alloys.

2. Experimental Procedure

2.1 Process and materials

Experimental ingots were prepared by vacuum induction melting (VIM) and then refined using vacuum arc remelting (VAR). All of the ingots were rolled to Φ18 mm bars after homogenizing heat treated at 1423 K for 20 h, then the temperature was increased to 1453 K for 48 h, followed by air cooling. Each ingot was about 50 kg in mass and the chemical compositions of the resulting alloys are listed in Table 1. The test samples from Φ18 mm bars were each solution treated at 1233 K for 1 h, air cooled and aged at 1061 K for 8 h, furnace cooled at 329 K/h to 977 K for 8 h and then air cooled. Several different ingots were prepared in which the Al+Ti ratios were varied, set at, the 3.9%, 4.3%, 4.8%. In addition, the atomic percent ratio of the Al and Ti was varied in these ingots with Al/Ti(at%) ratios of 3.0, 3.5 and 4.0.

2.2 APT analysis

The distribution of aluminum in alloy 2 was examined using the APT technique. Prior to the APT characterization...
The samples were heat treated as follows: 1233 K for 1 h, air cooled, aged at 1061 K for 8 h, furnace cooled at 329 K to 977 K for 8 h, air cooled. All the samples were cut into square rods with a size of 20 mm × 0.5 mm × 0.5 mm and then were electropolished to sharp needle shape specimens using a two-step procedure. The first step employed an electrolyte composed of 25% perchloric acid and 75% acetic acid with a direct current of 15 V at room temperature and the second step employed an electrolyte composed of 4% perchloric acid in 2-butoxethynal at 20 V. A LEAP 3000 HR APT was employed for the microstructure and microanalysis investigations. The largest analysis area of the APT was 150 nm × 150 nm and selected analysis area was 60 nm × 60 nm.

### 2.3 Phase Identification

Quantitative determination of the different phases of the three materials was further investigated by electrolytic phase separation followed by micro-chemical analysis. The procedure of phase extraction and separation was illustrated in Table 2 and the detailed phase identification was performed using the SEM and XRD techniques.

### 3. Results and Discussions

#### 3.1 Results

**3.1.1 Microstructures**

The microstructures of the three materials are shown in Fig. 1 and the average grain sizes can be seen to be about ASTM 11.

![Fig. 1 Microstructures of three samples; (a) Alloy 1, (b) Alloy 2, (c) Alloy 3.](image-url)

The detailed microstructure was observed using a field emission scanning electron microscope, SUPRA 55. It was found that there was little δ-phase in the three as-rolled materials. Following solution heat treatment, a large amount of γ′ and δ phases appeared to precipitate, as shown in Fig. 2. There is no significant difference in the morphology of δ phase in the three alloys. It appeared to primarily distribute along the grain boundaries with a lath-shaped exhibiting some angles between the δ phase and grain boundary.

![Fig. 2 Microstructures of three samples; (a) Alloy 1, (b) Alloy 2, (c) Alloy 3.](image-url)

### Table 1 Chemistry of experimental alloys.

<table>
<thead>
<tr>
<th>Alloy No.</th>
<th>Ni</th>
<th>Co</th>
<th>Fe</th>
<th>Cr</th>
<th>Al</th>
<th>Mo</th>
<th>Nb</th>
<th>Ti</th>
<th>W</th>
<th>C</th>
<th>P</th>
<th>B</th>
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</thead>
<tbody>
<tr>
<td>1</td>
<td>51.27</td>
<td>9.01</td>
<td>9.20</td>
<td>18.97</td>
<td>1.36</td>
<td>2.78</td>
<td>5.52</td>
<td>0.75</td>
<td>1.08</td>
<td>0.041</td>
<td>0.013</td>
<td>0.0051</td>
</tr>
<tr>
<td>2</td>
<td>50.98</td>
<td>8.99</td>
<td>9.28</td>
<td>18.98</td>
<td>1.56</td>
<td>2.78</td>
<td>5.55</td>
<td>0.75</td>
<td>1.07</td>
<td>0.044</td>
<td>0.012</td>
<td>0.0062</td>
</tr>
<tr>
<td>3</td>
<td>50.62</td>
<td>8.98</td>
<td>9.29</td>
<td>19.12</td>
<td>1.78</td>
<td>2.81</td>
<td>5.51</td>
<td>0.75</td>
<td>1.08</td>
<td>0.039</td>
<td>0.012</td>
<td>0.0055</td>
</tr>
</tbody>
</table>

### Table 2 Procedures of phase extraction and separation.

<table>
<thead>
<tr>
<th>Procedures</th>
<th>Solutions</th>
<th>Parameters</th>
</tr>
</thead>
<tbody>
<tr>
<td>1</td>
<td>1%(NH₄)₂SO₄+2% Citric acid+H₂O</td>
<td>T = 5<del>10°C, t = 1 h, i = 0.02</del>0.025 A/cm²</td>
</tr>
<tr>
<td>2</td>
<td>3.6%ZnCl₂+5%HCl+1%Tartaric acid+Methanol</td>
<td>T = −5°C, t = 1~2 h, i = 0.10 A/cm²</td>
</tr>
<tr>
<td>3</td>
<td>4%Sulfosalicylic acid+1%LiCl+5% Glycerine+Methanol</td>
<td>T = −10~−7°C, t = 1~2 h, i = 0.1 A/cm²</td>
</tr>
</tbody>
</table>

<table>
<thead>
<tr>
<th>Procedures</th>
<th>Solutions</th>
<th>Parameters</th>
</tr>
</thead>
<tbody>
<tr>
<td>1</td>
<td>200ccH₂SO₄+200ccH₂O+20gTartaric acid</td>
<td>Reflux for t = 2~3 h, δ + α-Cr dissolved</td>
</tr>
<tr>
<td>2</td>
<td>5%H₂SO₄+7%Tartaric acid+H₂O</td>
<td>Boil bath for t = 3 h, γ′ + γ″ dissolved</td>
</tr>
</tbody>
</table>
precipitation was depleted around the δ phase which is considered to be beneficial to plastic deformation of grain boundaries. But it was noted that the amount of γ’ precipitation increased significantly with increasing aluminum content.

Following aging heat treatment, the finer sized γ’ continued to precipitate and the γ’ precipitation clearly exhibited a spherical morphology (in Fig. 3). It would appear that during air cooling and solution heat treatment, coarse sized γ’ and δ phases precipitate more than the finer sized γ’ does during aging heat treatment. Therefore, the distribution of γ’ can be classified into primary and secondary γ’. The primary γ’ was believed to precipitate mostly during air cooling for rolled-bar since the solution temperature of 1233 K is below the solvus temperature of γ’, so that the primary γ’ cannot completely dissolve into the matrix. Therefore the observed primary γ’ should be the primary γ’ that remained after the solution heat treatment. In addition, the primary γ’ can also precipitate during air cooling from heat-rolled or high temperature solution. Very little primary γ’ was observed when the aluminum content was less than 2.9 at% (1.35 mass%). From discussion, it can be concluded that the primary γ’ illustrated in Fig. 3 can precipitate during air cooling after being rolled or in a high temperature solution and the secondary γ’ precipitates during aging heat treatment.

3.1.2 Tensile property

All the experimental materials were subjected to a standard heat treatment at 1233 K for 1 h, air cooled, and aged at 1061 K for 8 h, furnace cooled at 329 K/h to 977 K for 8 h. The tensile property of three materials with different aluminum contents is shown in Fig. 4 and the experimental data was the average of three experiments. These data reveal that as the aluminum content of the alloy was increased in increments of 0.45 at% (0.2 mass%) the ultimate tensile and yield strength at 977 K increased to approximately 40 MPa and 60~90 MPa with each successive aluminum incremental increase. In addition, these same incremental increases reduced the area and alloy elongation reduced by about 8% and 4~6% per 0.45 at% (0.2 mass%) aluminum increase. In addition, the ultimate tensile strength at 923 K was almost the same as the parent alloy and yield strength increased by about 40 MPa, and reduction of area and elongation reduced by about 5~23% and 6~10% respectively with increasing per 0.45 at% (0.2 mass%) Al.

The tensile fracture morphology of the experimental ingots is shown in Fig. 5 which was typical transgranular with dimple fractures for three samples. But the ductile fracture area decreases gradually, and dimple depth became shallower as the aluminum content of the alloy increased.

3.1.3 Modeling

The effect of aluminum on phase precipitation was predicted using JMatPro 6.0 software. The quantitative prediction results are listed in Table 3 and indicated that the amount of γ’ increases and the δ decreases gradually, while the solvus temperature of γ’ rises as the δ temperature gradually decreases with the increase in the aluminum content of the alloy. With increases of 0.45 at% (0.2 mass%) Al, it was observed that the γ’ phase increases at 3 mass% Al with a decrease of the δ phase at 1 mass% Al. In addition, the solvus temperature of the γ’ phase rises by about 288 K (15°C) accompanied by a 281 K (8°C) decrease in solvus temperature of the δ phase. Therefore, aluminum appears to have a significant effect on the phase precipitation of 718Plus. In practical hot working, the effect of varying the aluminum content on phase precipitation should be considered.

Figure 6 shows that the plot of the equilibrium content of δ and γ’ phases at 973 K versus the amount of aluminum produced a straight a line. So the equilibrium content of the δ and γ’ phases at 973 K is proportionate to the aluminum weight percent in the range of 2.4 at% (1.1 mass%) to 4.1 at% (1.9 mass%). It was also found that the content of the δ and γ’ phases can be approximately described by the relationship $Y(δ) = −5.5X + 17.6$ and $Y(γ’) = 14.4X − 2.4$, where X denotes the aluminum content in weight percent. Little
Fig. 4  Tensile properties of three materials with increasing aluminum.

Fig. 5  The tensile fracture morphology of three alloys; (a) Alloy 1, (b) Alloy 2, (c) Alloy 3.

Table 3  The solvus temperature and equilibrium content of γ' and δ phase of three alloys.

<table>
<thead>
<tr>
<th>Alloy No.</th>
<th>Solvus temperature of γ' (K)</th>
<th>Equilibrium content of γ' at 973 K (mass%)</th>
<th>Solvus temperature of δ (K)</th>
<th>Equilibrium content of δ at 1233 K (mass%)</th>
</tr>
</thead>
<tbody>
<tr>
<td>1</td>
<td>1245</td>
<td>17.1</td>
<td>1299</td>
<td>6.2</td>
</tr>
<tr>
<td>2</td>
<td>1260</td>
<td>20.0</td>
<td>1291</td>
<td>5.3</td>
</tr>
<tr>
<td>3</td>
<td>1272</td>
<td>23.0</td>
<td>1283</td>
<td>4.3</td>
</tr>
</tbody>
</table>
M₃B₂ and M₂₃C₆ are predicted to precipitate in alloy 718Plus. It is noted that the σ-phase was predicated to precipitate in 718Plus and the amount of σ-phase will increase with increasing aluminum content. This is consistent with the experimental results illustrated in Fig. 3. As with the TCP phase, the σ-phase is considered to be harmful to mechanical properties. But the effect of the σ-phase on mechanical properties of 718Plus and the quantitative changes during long aging times requires further investigation.

3.1.4 APT results

As can be seen in Fig. 7, it is evident that the concentration of Nb in γ' phase is higher than in matrix and most of the Al is distributed in γ' phase. Moreover, it was obvious that the concentration of Al and Nb in δ phase was about 12.5 at% (6.0 mass%) and 8.5 at% (13.8 mass%) and in the δ phase these concentrations were 6 at% (2.7 mass%) and 13 at% (2.4 mass%). At the δ/matrix interface, the highest concentration of Al and Nb was present at 11% (5.0 mass%) and 15 at% (22.5 mass%), respectively. Therefore, it was concluded that Al is primarily distributed in γ' and δ phase and Nb is also present in the matrix but is primarily distributed in the γ' and δ phases.

3.1.5 Phase identification

Table 4 shows the chemical compositions of δ and γ' phases in atom percent and the weight percent of in the three materials. These data show that the quantity of γ' in these three samples was 22.3%, 24.4% and 27.4% and the quantity of the δ phase in these samples was 2.0%, 1.5% and 0.7%. Therefore, the amount of γ' phase increased, while the quantity of the δ phase decreased gradually with increasing aluminum in the alloy. The quantity of Cr, Fe, Nb, Ti, W decreased as the amount of Al and Ni increased slightly in γ', while the amount of Co, Cr, Ti decreased as Al content increased in the δ phase. The empirical formulae for the δ phase can be depicted as \( (\text{Ni}_{0.90}\text{Co}_{0.08}\text{Fe}_{0.01}\text{Cr}_{0.01})_{3} (\text{Cr}_{0.02}\text{Al}_{0.17}\text{Mo}_{0.02}\text{Nb}_{0.58}\text{Ti}_{0.20}\text{W}_{0.01}) \), \( (\text{Ni}_{0.90}\text{Co}_{0.08}\text{Fe}_{0.01}\text{Cr}_{0.01})_{3} \).

Table 4 Chemical compositions of constituent phases in three examples.

<table>
<thead>
<tr>
<th>Alloy No.</th>
<th>Total Percent (mass%)</th>
<th>Compositions (at%)</th>
<th>δ phase</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>Ni</td>
<td>Co</td>
<td>Fe</td>
</tr>
<tr>
<td>1</td>
<td>67.07</td>
<td>6.37</td>
<td>0.89</td>
</tr>
<tr>
<td>2</td>
<td>67.33</td>
<td>6.22</td>
<td>0.78</td>
</tr>
<tr>
<td>3</td>
<td>67.49</td>
<td>5.85</td>
<td>1.03</td>
</tr>
<tr>
<td></td>
<td>Ni</td>
<td>Co</td>
<td>Fe</td>
</tr>
<tr>
<td>1</td>
<td>22.344</td>
<td>68.30</td>
<td>2.98</td>
</tr>
<tr>
<td>2</td>
<td>24.399</td>
<td>68.60</td>
<td>3.02</td>
</tr>
<tr>
<td>3</td>
<td>27.396</td>
<td>69.14</td>
<td>2.96</td>
</tr>
</tbody>
</table>
(Cr_{0.02}Al_{0.22}Mo_{0.02}Nb_{0.52}Ti_{0.20}W_{0.01}) and (Ni_{0.90}Co_{0.06-Fe_{0.01}}Cr_{0.01}) (Cr_{0.02}Al_{0.22}Mo_{0.02}Nb_{0.53}Ti_{0.18}W_{0.01}) in three materials. The empirical formulae for the γ' phase can be depicted as (Ni_{0.09}Co_{0.04}Fe_{0.02}Cr_{0.01}) (Cr_{0.02}Al_{0.02}Mo_{0.02-Nb_{0.32}Ti_{0.18}}W_{0.01}), (Ni_{0.09}Co_{0.04}Fe_{0.02}Cr_{0.01}) (Cr_{0.02}Al_{0.44}Mo_{0.02-Nb_{0.34}Ti_{0.12}}W_{0.01}) and (Ni_{0.02}Co_{0.04}Fe_{0.02}Cr_{0.02}Ni_{0.07}Al_{0.40}Mo_{0.01}Nb_{0.33}Ti_{0.12}W_{0.01}) in three materials. Experimental results indicated that the true lattice structure of the δ phase in 718Plus alloy was consistent with hexagonal η-Ni3Ti structure, but its chemistry was close to Ni3Al_{0.5}Nb_{0.5}, which was also reported by Xie et al. Pickering et al. considered that the δ phase in 718Plus alloy was consistent with η-Ni3Ti, but its chemical composition was close to Ni_{6}AlNb with partial ordering of Al and Nb over the prototype Ti sites. Actually, Ni3Al_{0.5}Nb_{0.5} and Ni_{6}AlNb have identical chemical composition, but the Ni_{6}AlNb structure may be accurate, because the structure of Ni3Al_{0.5-Nb_{0.5}}, first reported in the 1960s by Gieseen et al., had the Al and Nb atoms randomly distributed over the Ti sites of the η-Ni3Ti crystal structure. The experimental demonstrate that Nb, Ti and Al atoms are ordered over the two distinct Ti sites of the η-Ni3Ti prototype structure in 718Plus alloy. It is suggested that this new phase be named the δ'-phase to distinguish it from the true η-Ni3Ti and δ-Ni_{3}Nb phases.

The size distribution of γ' phase in three materials was analyzed as can be seen in Fig. 8. The size of the γ' is not continuously distributed, most of the γ' is distributed at 10~36 nm, with few greater than 60 nm. The experimental results shown in Fig. 3 indicate that γ' phase greater than 60 nm precipitate during air cooling and solution heat treatment which can be regarded as the primary γ'. After the heat aging treatment, a large number of the 10~36 nm γ' precipitate and this portion of the γ' can be regarded as the secondary γ'. With increasing aluminum, the amount of 10~36 nm γ' precipitate and this portion of the γ' can be regarded as the secondary γ'. Because the γ' phase predominately precipitates at the grain boundary and δ-Ni_{3}Nb phases.

3.2 Discussion

With the exception of different aluminum contents, the chemical compositions and grain size of three materials were almost the same. Therefore, the variation of properties is probably due mainly to the difference of the aluminum content. From the APT results, aluminum appeared to be mainly situated in γ' and δ phases. As the aluminum was increased from 2.9 to 3.8 at% (1.35 to 1.75 mass%), the amount of total γ' increased significantly from 22.3% to 27.4 mass%, while the δ phase decreased from 2.0% to 7.7 mass%. The modeling and experimental results were consistent with each other. The γ' phase is the strengthening phase in alloy 718Plus and the improvement in strength with increasing aluminum is related primarily to the increase of the total γ' content.

As is well known, δ precipitation is important for adjusting the properties in alloy 718. Although the precise effect of the δ phase can vary considerably in the 718 alloy and it appears to depend strongly on the location and morphology of this phase. It is generally accepted that a small amount of δ precipitation is essential for grain boundary pinning, which can lead to serration of grain boundaries in 718 and reports suggest that such a microstructure is likely to improve intergranular fatigue crack resistance. But, excessive levels of the coarse δ are detrimental to properties due to solute depletion of the γ' phase in 718. Although δ precipitation in 718Plus proved to actually be η-Ni3Al_{0.5-Nb_{0.5}}, its morphology and chemical composition indicated that it should have an effect similar to the δ phase in 718. Because of its morphology, the δ-phase does not contribute significantly to the hardening of the alloy. On the contrary, its presence implies a loss of hardenability due to the depletion of γ', which can be seen from Fig. 2. But moderate fractions of δ are effective in pinning grain growth during solution treatment and the appropriate morphology has been shown to provide resistance to grain boundary creep fracture. It can restrict the slip of grain boundaries and cracking path, which are beneficial for relieving stress concentration and eliminating notch sensitivity. The beneficial effect observed by Pickering et al. is the δ phase in 718Plus influences the morphology of grain boundaries such that the latter appeared distorted or serrated. The fracture morphology illustrated in Fig. 5 shows that the ductile fracture area decreased gradually from macrofracture morphology and the dimple depth became shallower with increasing aluminum content. All of the fracture morphology was typically transgranular and as was the dimple fracture for the three alloys. Therefore, it can be inferred from these results that the decrease of ductility was related to the increase of γ' content in the matrix. As discussed above, an increase of Al content caused the amount of δ phase to decrease. Because the δ phase can restrict the slipage of grain boundaries and cracking paths, both of which are beneficial for relieving stress concentration and eliminating notch sensitivity. So the decrease of ductility was also related to the decrease of the δ phase at grain boundary.
718, due to the slow diffusion rate of Nb. From Table 3, the solvus temperature of $\gamma'$ rises while the $\delta$ phase falls with increasing aluminum. The change of solvus temperature was considered to be related to the variety of chemical compositions of $\gamma'$ and $\delta$ phases in three alloys. This can be seen from Table 4 that the as the Al content increased the amount of Cr, Fe, Nb, Ti, W in the alloy decreased while the Al and Ni content increased slightly in $\gamma'$ phase. Simultaneously, as the Al content was increased in the alloy, the amount of Co, Cr, Ti in the alloy decreased and Al increased slightly in the matrix, so the change of solvus temperature of alloy 718Plus is mainly related to the variety of chemical compositions of $\gamma'$ and $\delta$ phases in the alloy. The change of solvus temperature of $\gamma'$ phase was considered to be related to the variety of chemical compositions of $\gamma'$ and $\delta$ phases in three alloys. This can be seen from Table 4 that the as the Al content increased the amount of Cr, Fe, Nb, Ti, W in the alloy decreased while the Al and Ni content increased slightly in $\gamma'$ phase. Simultaneously, as the Al content was increased in the alloy, the amount of Co, Cr, Ti in the alloy decreased and Al increased slightly in the $\delta$ phase. In practical hot working, the effect of different Al content on phase precipitation should be considered.

From above discussion, it can be concluded that the content of aluminum can significantly influence the quantity, solvus temperature and the chemical compositions of the $\gamma'$ and $\delta$ phases which promotes a gradually changes of the mechanical properties of the 718Plus alloy.

4. Conclusions

(1) There is a significant influence from the addition of aluminum on the amount, solvus temperature and chemical compositions of the $\gamma'$ and $\delta$ phases in alloy 718Plus. As the aluminum content is increased to 3.83 at% (1.75 mass%), the amount of $\gamma'$ increased from 22.3% to 27.4 mass% and the amount of the $\delta$ phase decreased from 2.0% to 0.7 mass% after a standard heat treatment. The chemical composition of the $\gamma'$ and $\delta$ phases also changed with increasing aluminum which leads to a rise in the solvus temperature of $\gamma'$ phase and a decrease in the solvus temperature of the $\delta$ phase as the aluminum content is increased by increments of 0.45 at% (0.2 mass%) Al.

(2) The size of $\gamma'$ phase in alloy 718Plus can be classified into the primary and secondary $\gamma'$. The primary $\gamma'$ was greater than 60 nm which precipitates mainly during air cooling for rolled-bar and solution heat treatment, and the secondary $\gamma'$ was 10~36 nm in size with contents as high as 90% in total $\gamma'$, precipitate during heat treatment. The true $\delta$ phase in 718Plus was confirmed as $\eta$-Ni$_3$Al$_{0.3}$Nb$_{0.5}$, which has $\eta$-Ni$_3$Ti structure, but with Ni$_3$Al$_{0.3}$Nb$_{0.5}$ chemistry.

(3) The mechanical properties testing showed that the tensile yield strength improves, but ductility falls gradually with the increasing of Al content. It approved that the improvement of strength for alloys was related to the increasing of the amount of $\gamma'$ phase and the decrease of ductility was mainly related to increasing $\gamma'$ content in the matrix, but the decrease of $\delta$ phase at grain boundary may also have played an important role.

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