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It will be published in the Johnson Matthey Technology Review

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Research Progress of Pt-Base Superalloys for High Temperature Applications

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Abstract
Pt-based alloys are being developed for high-temperature applications with the aim of replacing some of the currently used Ni-based superalloys (NBSAs) and benchmark alloy, PM2000. The Pt-based superalloys have a similar structure to the NBSAs and can potentially be used at higher temperatures and in more aggressive environments because Pt is more chemically inert and has a higher melting point. In this paper, the recent progress in research and development of Pt-based superalloys is overviewed. Firstly, the composition optimization and structural design of Pt-base superalloys are introduced. The structural characteristics, mechanical properties, oxidation resistance and corrosion behavior of Pt-Al ternary, quaternary and multiple superalloys are summarized. Finally, directions for further research and application of Pt-base superalloys are analyzed and prospected.

Introduction
Components in many different applications at high temperature and in corrosive environments require materials with excellent high temperature mechanical properties and chemical resistance. Aerospace applications represent an extremely challenging field, for the development of new materials, and for the improvement of the existing ones (1). NBSAs with the Ni3Al intermetallic compounds as the main strengthening phase have been widely used in high-temperature components such as aero engines, industrial gas turbine blades and modern industry fields. After decades of development, the working temperature of NBSAs is about 1100ºC (2) and can reach up to 1150ºC (3, 4). Recently, a new Pt-modified Ni-base alloy with exceptional high temperature stability has been identified (5, 6). Coarsening studies conducted reveal unusually high volume fractions of morphologically stable γ' precipitates up to 1200°C, which suggests that the alloy would have excellent performance as a bond coat and/or single crystal blade. A further increase in the operating temperature of the gas turbine
will improve the combustion efficiency, reduce fuel consumption and CO\textsubscript{2} emissions, and leads to higher thrust values (7, 8). NBSAs operating temperature is approximately 85% of their melting point. An increasing interest has been shown in developing new alloys based on materials with higher melting points with similar structure to that of NBSAs and capable of being used at 1300°C (9, 10).

Potential candidate materials that may replace NBSAs or Fe-based superalloys such as PM2000 mainly include intermetallic compounds, ceramics and ceramic matrix composites, refractory metals and high melting point platinum group metals (11). Some high melting point intermetallic compounds have high-temperature strength, lower diffusion rate and excellent corrosion resistance. However, they lack plasticity or fracture toughness at room temperature (12, 13). Components based on ceramics and ceramic matrix composites have high-temperature strength, creep resistance, oxidation resistance and corrosion resistance which are attractive for their potential use in gas turbine engines. Unfortunately, low fracture toughness and brittle behaviour usually associated with ceramics are problems for high-temperature applications. Refractory metals and their alloys have high melting temperatures and for this reason researchers are considering the possibility of using them in the hot parts of gas turbines to replace NBSAs. These refractory metals and alloys lack sufficient oxidation resistance which limits their practical application (14). Superalloys based on platinum group metals (Pt, Ir, Rh) show extremely strong chemical stability and two-phase structure (fcc/L1\textsubscript{2}), which makes this group of alloys a potential candidate to be developed as high-temperature materials for next-generation gas turbines (15-22). However, the main weaknesses of most Ir-based and Rh-based refractory superalloys are brittleness, high cost and high density.

Unlike iridium and rhodium, platinum has become an essential high-temperature material in special applications. It has a higher melting point than Ni (Pt = 1769°C, Ni = 1453°C), better oxidation resistance, corrosion resistance, chemical stability and does not require coating protection when used at high temperatures (23, 24). Pt-based alloys have excellent mechanical properties such as high creep strength and ductility, which make them have application potential in the fields of chemical engineering, space technology and glass industry (25). At this time, the research on Pt-based superalloys includes solid solution strengthened, dispersion strengthened and precipitation strengthened alloys as well as platinum group metal compounds (26). Current usage is restricted to solid solution strengthened alloys and dispersion strengthened alloys, the latter being classified as part of the group of composite materials. Solid solution strengthened
Pt-based alloys are a family of alloys which have been researched and developed for some time. The compositions and preparation process can be said to be more mature.

All transition group elements have considerable solid solubility in Pt. The elements near Pt in the periodic table form a continuous solid solution with Pt and have different degrees of solid solution strengthening effect on the Pt matrix. At high temperatures, Ru, Ir, Rh have higher tensile strength than Pt and Pd. The high temperature durability and creep rupture strength of Ir are also much higher than that of Pt-Rh alloys. Ru, Ir, Rh, Pd have become the main solid solution strengthening elements. According to the relationship between stacking fault energy and creep rate, Ru and Ir have the largest solid solution strengthening effect on Pt, followed by Rh, and Pd with the smallest effect (27).

Currently, the research and development on Pt-based solid solution alloys mainly include binary alloys such as Pt-Rh, Pt-Ir, Pt-Ru, Pt-Ni, Pt-W, and ternary alloys such as Pt-Pd-Rh and Pt-Rh-Ru (28). The properties of Pt-Rh alloys are the most stable: an increase in Rh content leads to higher temperature durability, extended creep life and decreased creep rate (29). However, the improvement of mechanical properties decreases when Rh content exceeds 30 wt%. In addition, machinability is significantly worse at these levels of Rh content. The high temperature durability, creep life and creep rate of Pt-Ir alloy are better than the Pt-Rh alloy. On the other hand, Pt-Ir alloys tend to have higher weight losses in oxidizing environments above 1100°C due to the selective oxidation of Ir after prolonged exposure to such atmospheres. In addition, for solid-solution strengthened Pt-Rh alloys, the coarsening of crystal grains at high temperatures will lead to reduced alloy strength and premature failure of components.

In order to improve the high-temperature mechanical properties of Pt-based alloys, oxide dispersion strengthened (ODS) alloys with ZrO₂ or Y₂O₃ as reinforcing phase have been researched and developed (30, 31). The fine oxide particles dispersed in the Pt matrix can stabilize the grain boundaries, prevent the movement of dislocations and improve the high temperature fracture strength. However, these ODS alloys have great brittleness, crack sensitivity and cannot withstand severe temperature changes. A particular type of ODS alloys has been developed, namely dispersion hardened platinum (DPH) alloys (32). These alloys are reinforced by dispersion of oxides formed inside the alloy via an internal oxidation process of pure oxygen-reactive elements (i.e., elements with high affinity to oxygen such as Ce, Y, Sc and Zr). These elements are added to the melt and their oxidation is subsequently obtained by appropriate treatments. The stress-fracture strength of DPH alloys is further improved with respect to solution
strengthened alloys. For instance, the stress-fracture strength of DPH Pt is higher than that of Pt-10Rh alloy (33), with good plasticity. However, the production process of all the dispersion strengthened alloy families is complicated and problems may arise during welding operations. The strength of the welded joint tends to drop due to excessive formation of oxide particles during welding (24).

Precipitation strengthening is the main strengthening mechanism of Pt-based superalloys currently under research and development. These alloys show higher temperature strength than solution strengthened and dispersion strengthened alloys at 1300ºC (34–39). However, high density and cost are the major drawbacks to the use of Pt, but it is likely that the Pt-based alloys can be used for the highest application temperature components (10, 40). Due to the excellent properties of oxidation and corrosion resistance, the Pt-Al system superalloys could have potential as coatings on NBSAs or other substrates (10, 22, 41). Their density can be slightly reduced by adding suitable light alloying elements, while the high performance and recyclability of Pt-based alloys make up for the high price (42). This paper summarizes the research status and progress of Pt-based superalloy materials. Firstly, we introduce the composition and structural optimization design of Pt-based superalloys, the structural characteristics and evolution of Pt-Al-based ternary, quaternary and multi-element superalloys, and their mechanical properties, oxidation and corrosion resistance behaviors. The strengthening mechanisms, the relationship between oxidation, corrosion and alloy composition have been analyzed and the results will be presented and discussed. Finally, further research and application prospects of Pt-based superalloys are analyzed and discussed.

1 Structure and composition design of Pt-based superalloys

A large number of new alloys have been researched and developed based on strong demand for higher working temperature and high temperature resistant structural materials in the aerospace field. The goal of the research is to seek a material with better high-temperature mechanical properties (tensile, fracture, creep and thermo-mechanical fatigue properties) and environmental stability (resistance to high-temperature oxidation and hot corrosion) than Ni-based superalloys (43). Inspired by the successful experience of precipitation strengthening obtained in the γ matrix (fcc structure) of Ni-based superalloys, much effort is now put into the research and development of Pt-based superalloys with similar structures to the γ/γ' system found in Ni-based superalloys (7). Research on Pt-based superalloys began at the end of the 20th century. The initial work was mainly on structure and composition design, including the formation of
a Pt\textsubscript{3}X (γ') precipitation strengthening phase and the selection of solid solution strengthening elements.

### 1.1 Second phases and possible reinforcement

Pt can form Pt\textsubscript{3}X and Pt\textsubscript{5}X high melting point intermetallic compounds on the Pt-rich side of the phase diagram with transition metals and rare earth metals (44), such as Pt\textsubscript{3}Al (1550ºC), Pt\textsubscript{3}Hf (2250ºC), Pt\textsubscript{3}Sc (1850ºC), Pt\textsubscript{3}Y (2020ºC), Pt\textsubscript{3}Zr (2250ºC). Their melting points are higher than Ni\textsubscript{3}Al (1390ºC). Most Pt compound precipitation phases with γ' structure have a high melting point, high thermal conductivity, low thermal expansion coefficient, high strength and a large number of possible slip systems. It can be expected that γ/γ' Pt alloys will have high thermal strength and precipitation strengthening effects, leading to the possible development of a new generation of precipitation strengthened Pt-based superalloys (45). Figure 1 represents the binary Pt–Al phase diagram (44). The diagram shows that the maximum solubility of Al in Pt is about 10 at. % at relatively low temperatures, whereas at the eutectic temperature (1507ºC) the value is slightly higher. The Pt\textsubscript{3}Al phase forms at the eutectic temperature and the eutectic reaction can be found in the top right of the diagram. At high temperatures, the Pt\textsubscript{3}Al intermetallic compound shows a L1\textsubscript{2} structure. It transforms into a tetragonal structure (D\textsubscript{0}h) at lower temperatures. Pt\textsubscript{3}Al is the most important intermetallic phase.

![Fig.1. Binary Pt–Al equilibrium phase diagram (44)](image)

Table 1 lists the common precipitation phases and main performance evaluations in Pt-based alloys (46). The Pt\textsubscript{3}X phase mainly appears in Pt-transition metals and Pt-simple metal alloy systems, while Pt\textsubscript{5}X mainly appears in Pt-rare earth and Pt-alkaline earth metal alloy systems. The Pt\textsubscript{3}X phase formed by Cr, V
and Pt decomposes at 1130°C and 1015°C respectively, and was excluded from the design study of Pt-based superalloys. Although Sn, Pb, Ga and other elements can form a L1₂ structure phase with Pt, they are also excluded due to the low melting point. Initial research on binary systems such as Pt-Zr, Pt-Hf and ternary system alloys such as Pt-Rh-Zr, Pt-Rh-Hf showed that γ' phases Pt₃Zr and Pt₃Hf are formed in the alloy which are coherent with the matrix (47). The presence of Zr and Hf as solid solution strengthening elements in combination with γ' precipitation strengthening give these alloys high-temperature mechanical properties, but they all have the problem of poor oxidation resistance. The formation of brittle Zr and Hf oxides leads to embrittlement of the material (48).

The refractory metals Ta and Nb can increase the solidus temperature of Pt-based superalloys and are worthy of further study. Alloying with Al leads to the formation of Pt₃Al, a strengthening intermetallic of the Pt₃X type that has two different crystal structures: the cubic structure (L1₂) at high temperatures and allotrope tetragonal structure (D024) at low temperatures. The high-temperature allotrope can be stabilized at lower temperatures by adjusting the composition of the alloy matrix.

It is necessary to study the structure and properties of the Pt₃X second phase particles in order to develop precipitation strengthened Pt-based superalloys. However, there are few reports on the experimental measurement of the mechanical properties of Pt₃X intermetallic compounds. Adjal (49), Pan (50) and Li (51) investigated the electronic structure, thermodynamic properties, oxidation resistance and mechanical properties of Pt₃X intermetallic compounds by simulation calculation methods. The mechanical properties and anisotropy of Pt₃M alloy were determined from these studies. The anisotropy is mainly derived from the d electronic state of Pt and the d electronic state (or p electronic state) of M. Pt₃Hf has the highest modulus (bulk modulus, shear modulus, Young's modulus) and hardness, while Pt₃Y has the lowest values (51). Liu (52) used first-principles calculations to study the effect of pressure on the structure and mechanical properties of Pt₃Al. The study found the elastic modulus, bulk modulus and shear modulus of Pt₃Al increase linearly with increasing pressure. Pt₃Al changes from tetragonal to cubic structure when the pressure reaches 60 MPa, indicating that the cubic structure of Pt₃Al has higher resistance to volume deformation.

### 1.2 Alloy composition design

Table 2 lists the candidate solid solution strengthening elements and main performance evaluations of Pt-based superalloys (46). Pt group metals Ru, Ir, Rh, Pd and Ni, Re can all be used as solid solution strengthening elements. The alloys Pt-Al-Ni, Pt-Al-Ru, Pt-Al-Cr, Pt-Al-Ti, Pt-Al-Re, Pt-Ti-Ru, Pt-Ti-Re, Pt-Ta-Ru, Pt-
Ta-Re and Pt-Nb-Ru were selected through the comprehensive evaluation of precipitation strengthening, solid solution strengthening and alloy properties. These alloys were optimized and screened and are all characterized by a fcc (Pt) solid solution matrix and fcc (L1$_2$) Pt$_3$X precipitation phase, thus a two-phase microstructure. Investigations on structural composition, mechanical properties and oxidation resistance were performed. The results showed that the two-phase structure of Pt-Al-X and Pt-Ti-X alloys confers significant precipitation strengthening, with hardness exceeding 400 HV1 and strong resistance to crack initiation and propagation. Al-containing alloys also have better oxidation resistance than the other alloys due to the formation of an Al oxide-based protective film on the surface. Internal oxidation was observed in Ti-containing alloys. Therefore, Al is considered to be a necessary alloying element for the development of oxidation-resistant Pt-based superalloys (46, 53). For these reasons, the subsequent research mainly focuses on the Pt-Al-X alloy system.
<table>
<thead>
<tr>
<th>Element</th>
<th>Structure of Pt₃X</th>
<th>Melting range, °C</th>
<th>Environmental resistance</th>
<th>Density, g/cm³</th>
<th>Attributes</th>
</tr>
</thead>
<tbody>
<tr>
<td>Al</td>
<td>Low temperature: Tetragonal (D024)</td>
<td>1500–1769</td>
<td>Good oxidation resistance. Forms stable external Al₃O₃ scale which protect the metal from internal oxidation</td>
<td>2.7</td>
<td>Low density. Transformation of Pt₃Al at 1290°C. Good oxidation resistance. Lowers solidus temperature</td>
</tr>
<tr>
<td></td>
<td>High-temperature: Face-centred cubic (L1₂)</td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>Ti</td>
<td>Tetragonal (D024) at stoichiometric compositions. Face-centred cubic (L1₂) in Pt-rich alloys</td>
<td>1769–1800</td>
<td>Prone to oxidation even at low temperatures</td>
<td>4.5</td>
<td>L1₂ structure. Favourable density. Increases solidus temperature</td>
</tr>
<tr>
<td>V</td>
<td>Face-centred cubic (L1₂)</td>
<td>1769–1805</td>
<td>V absorbs relatively large amounts of oxygen</td>
<td>5.8</td>
<td>The Pt₃V phase is only stable to 1015°C</td>
</tr>
<tr>
<td>Cr</td>
<td>Face-centred cubic (L1₂)</td>
<td>1769–1785</td>
<td>Cr has a beneficial effect on hot corrosion and oxidation resistance</td>
<td>7.19</td>
<td>The Pt₃Cr phase is only stable to 1130°C</td>
</tr>
<tr>
<td>Ga</td>
<td>Face-centred cubic (L1₂)</td>
<td>1373–1769</td>
<td>The effect of Ga additions on high temperature oxidation of alloys is uncertain</td>
<td>5.91</td>
<td>Low melting temperature</td>
</tr>
<tr>
<td>Y</td>
<td>Face centred cubic (L1₂). No two-phase (Pt)₃Pt₃Y region because of the Pt-Y phase</td>
<td>1615–1769</td>
<td>The effect of major Y additions on the environmental behaviour of alloys has not been documented</td>
<td>4.5</td>
<td>Highly reactive and difficult to process</td>
</tr>
<tr>
<td>Zr</td>
<td>Tetragonal (D024) at stoichiometric compositions. Face-centred cubic (L1₂) in Pt-rich alloys</td>
<td>1769–1963</td>
<td>Exposure to oxygen causes embrittlement due to the formation of brittle oxides</td>
<td>6.4</td>
<td>L1₂ structure. Generally embrittling in alloys. Susceptible to oxidation</td>
</tr>
<tr>
<td>Nb</td>
<td>&lt; ~ 1100°C: orthorhombic</td>
<td>1769–2000</td>
<td>Oxidizes substantially at T &gt;500°C</td>
<td>8.55</td>
<td>Raises solidus temperature. Probably only partially coherent Pt₃Nb</td>
</tr>
<tr>
<td></td>
<td>&gt; ~ 1100°C: tetragonal (D024)</td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>Sn</td>
<td>Face-centred cubic (L1₂)</td>
<td>1365–1769</td>
<td>The high-temperature oxidation behaviour of Sn-based alloys is unknown</td>
<td>7.29</td>
<td>Low melting point and is not suitable for high-temperature use</td>
</tr>
<tr>
<td>Hf</td>
<td>Not reported. Pt-rich Pt₃Hf with L1₂ structure has been reported</td>
<td>1769–2000</td>
<td>Exposure to oxygen causes embrittlement due to the formation of brittle oxides</td>
<td>13.1</td>
<td>L1₂ structure. Highly reactive and difficult to process</td>
</tr>
<tr>
<td>Ta</td>
<td>Monoclinic (L60)</td>
<td>1769–1970</td>
<td>Prone to rapid oxidation at T &gt;500°C</td>
<td>16.6</td>
<td>Raises solidus temperature. Good high-temperature mechanical properties</td>
</tr>
<tr>
<td>Pb</td>
<td>Face-centred cubic (L1₂)</td>
<td>915–1769</td>
<td>The high-temperature oxidation behaviour of Pb has not been extensively studied</td>
<td>11.3</td>
<td>Low melting temperature</td>
</tr>
<tr>
<td>Element</td>
<td>Melting range, °C</td>
<td>Environmental resistance</td>
<td>Density, g/cm³</td>
<td>Attributes</td>
<td></td>
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<tr>
<td>Ni</td>
<td>1455–2447</td>
<td>A dense, almost pore-free layer of NiO is formed on exposure to oxygen. The diffusion rate of Ni atoms through this layer is low and this limits oxidation</td>
<td>8.9</td>
<td>Similar electronic structure to Pt. Reportedly one of the strongest solid-solution strengtheners in Pt. Low cost. Low melting temperature</td>
<td></td>
</tr>
<tr>
<td>Ru</td>
<td>1769–2100</td>
<td>Forms volatile oxides above ~1100°C resulting in grain PGMs boundary embrittlement</td>
<td>12.2</td>
<td>Greater solid-solution strengthening effect than other because of its hexagonal structure. Reasonable cost</td>
<td></td>
</tr>
<tr>
<td>Rh</td>
<td>1769–1963</td>
<td>Comparable to Pt. Intercrystalline oxidation does occur but can be controlled by alloying</td>
<td>12.4</td>
<td>Extremely high cost (about four times that of Pt). Limited resources would not be able to sustain a large increase in demand</td>
<td></td>
</tr>
<tr>
<td>Pd</td>
<td>1555–1769</td>
<td>Highest vapourisation rate of all the PGMs in the presence of oxygen</td>
<td>12.0</td>
<td>Low melting point. High</td>
<td></td>
</tr>
<tr>
<td>Re</td>
<td>1769–2450</td>
<td>Pure Re is resistant to oxidation up to 1000°C. Above this cost temperature volatile oxides are formed. Re has been reported to improve the hot corrosion resistance of Ni-base superalloys</td>
<td>21.0</td>
<td>High</td>
<td></td>
</tr>
<tr>
<td>Ir</td>
<td>1769–2447</td>
<td>Forms volatile oxides &gt;1196°C. Above 1100°C Ir has a superior oxidation resistance to Ru</td>
<td>22.5</td>
<td>High cost. Limited resources would not be able to sustain a large increase in demand. Known to have excellent high-temperature mechanical properties</td>
<td></td>
</tr>
</tbody>
</table>
2 Pt-Al-X Ternary alloy
2.1 Structural characteristics of ternary alloys

In order to obtain an effective strengthening effect on Pt-based ternary alloys it is necessary to determine the low-temperature structure of the Pt$_3$Al phase in the representative binary alloy. Wolff (22) used an electric arc furnace (EAF) to smelt the Pt-12Al (at. %) alloy and the material was then subjected to 96 hours solution annealing and subsequent ageing at 1350°C. Figure 2 shows some structural features of the Pt-12Al alloy. It can be seen that Pt$_3$Al precipitation phase exhibits strong directional cubic alignment and the dispersed sub-micron nature of the precipitation phase exhibits a typical bimodal size distribution. This special-orientation geometrical configuration is a highly coherent phase which has low mismatch strain. The Pt$_3$Al-$\gamma'$ phase additionally has a low-temperature variant and a lath or twin structure.

The maximum volume fraction of the Pt$_3$Al phase is limited to about 30% due to the maximum solubility of Al in Pt. The volume ratio of the precipitation phase in Ni-based superalloys can be as high as 70–80%. Additional information about the microstructural features of Pt$_3$Al can be inferred from Figure 3 where a TEM bright-field image microstructure characteristic of the precipitate phase in an alloy with slightly higher Al content, namely Pt-14Al (at. %) alloy (54).

Fig.2. TEM bright-field image of cuboidal Pt$_3$Al precipitates in Pt-12Al alloy after solution annealing and subsequent ageing at 1350°C (22). The magnification is 10000 times
Fig. 3. Bright-field TEM image of a Pt₃Al precipitate in a Pt-14Al alloy. P = stacked plates or laths; M = Pt matrix. The arrows indicate individual platelets between stacked plates (54).

Adding transition metals such as Ni, Ti, Cr, Ru, Ir, Re, Ta, etc. to Pt-Al alloys can improve the strengthening effect of γ'(Pt₃Al) precipitates, thermal stability, solid solution strengthening of the matrix and overall properties. Several typical ternary alloy microstructures are shown in Figure 4. The two-phase structure of Pt₃Al and Pt solid solution of all alloys has been confirmed by X-ray diffraction (XRD) experiments (55). The alloys with nominal compositions (at.%) of Pt-14Al-3Re, Pt-14Al-4Ti, Pt-14Al-4Ta and Pt-14Al-4Cr were all characterized by a microstructure consisting of primary Pt₃Al surrounded by a fine two-phase eutectic-like mixture of a (Pt) matrix and fine particles of Pt₃Al. The proportion of primary Pt₃Al in Pt-14Al-4Ti, Pt-14Al-4Ta and Pt-14Al-4Cr alloys is between 40% and 50% and in Pt-14Al-3Re alloy it is about 25% according to optical microstructure analysis. The fine martensite-like lamellar structure was observed in Pt-14Al-3Re (Figure 4a) and Pt-22Al-2Ru (Figure 4e) alloys. This means the Pt₃Al phase has transformed from the high-temperature cubic structure (L1₂) to the low-temperature tetragonal structure (D0’c). In the Pt-14Al-4Ti and Pt-14Al-4Ta alloys, it seems that the high-temperature Pt₃Al phase is formed and maintained. This demonstrates that the third metal elements Ti and Ta can stabilize the L1₂ polymorph. It is important to observe that an alloying element must enter the Pt₃Al phase in order to prevent the low-temperature transformation of the Pt₃Al phase (55). Biggs (56) and Hill (57) have also shown the possibility for other alloy third components (Ni, Ti or Cr) in Pt-Al alloys to stabilize the cubic (L1₂) structure of Pt₃Al.
Fig. 4. SEM-BSE micrographs showing the microstructures of the Pt-Al-X alloys. (a) Pt-14Al-3Re; (b) Pt-14Al-4Ti; (c) Pt-14Al-4Ta; (d) Pt-14Al-4Cr; (e) Pt-22Al-2Ru (55)

Hill et al. (58) studied the microstructure and lattice mismatch of Pt-Al-X alloy systems with stabilizing elements Ti, Cr, Ta, Ru and Ir. Figure 5 shows TEM images of the typical two-phase microstructure. All the precipitated phases show a bimodal or even trimodal size distribution. Ti, Cr, Ta elements enter into the Pt$_3$Al phase to stabilize the cubic L1$_2$ structure, the precipitation phase has a cubic appearance with no clearly discernible internal structure (Figure 5(a)). On the contrary, when Ru and Ir enter the Pt matrix, the Pt$_3$Al precipitation phase transforms into a D0’c structure, and presents a band-like structure with alternating light and dark distributions (Figure 5(b)). Under higher magnification, it was confirmed that the lamellar structure in the precipitation phase of the D0’c tetragonal structure should belong to the twin structure (54).

Fig. 5. Typical TEM images of the ~ Pt$_3$Al precipitates in Pt-Al-X alloys (X = Cr, Ir, Ru, Ta or Ti), with letters indicating the different size ranges (where P = primary; I, T = intermediate; S = secondary): (a) L1$_2$ precipitates stabilized by Cr, Ta and Ti additions. The inset shows the selected area diffraction (SAD) pattern, confirming the L1$_2$ structure; (b) D0’c precipitates
The lattice misfits between the matrix and precipitate phase in the Pt-Al-X alloy system at room temperature and 800°C were measured respectively by XRD (58). The (220), (211) and (112) diffraction peaks are used to obtain the lattice constants of the Pt solid solution matrix, L12-Pt3Al and D0'c-Pt3Al precipitate phases, respectively (a_matrix and a_ppt represent the lattice constants of the Pt solid solution matrix and Pt3Al precipitation phase, respectively). Then, the Lattice misfits δ between the precipitation phase and the matrix are calculated by equation (1), and the results are listed in Table 3. It can be seen that the degrees of mismatch for all alloys are negative and the difference in mismatch degrees at different temperatures is very small. Besides alloys containing Ru, the lattice misfit increases at high temperature and the cubic L12 structure (alloys containing Cr, Ta or Ti) has a lower degree of mismatch than the D0'c structure (alloys containing Ir, Ru).

\[
\delta = 2 \left( \frac{(a_{\text{ppt}} - a_{\text{matrix}})}{(a_{\text{ppt}} + a_{\text{matrix}})} \right)
\]

Table 3 Lattice misfits between precipitates and matrix for selected Pt-Al-X ternary alloys (58)

<table>
<thead>
<tr>
<th>Alloy</th>
<th>Pt3Al type</th>
<th>Room temperature</th>
<th>800°C</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>a_matrix</td>
<td>a_ppt, nm</td>
<td>δ</td>
</tr>
<tr>
<td>Pt-10Al-4Cr</td>
<td>L12</td>
<td>3.9022</td>
<td>3.8741</td>
</tr>
<tr>
<td>Pt-10Al-4Ir</td>
<td>D0'c</td>
<td>3.8983</td>
<td>3.8507</td>
</tr>
<tr>
<td>Pt-10Al-4Ru</td>
<td>D0'c</td>
<td>3.9001</td>
<td>3.8530</td>
</tr>
<tr>
<td>Pt-10Al-4Ta</td>
<td>L12</td>
<td>3.8941</td>
<td>3.8682</td>
</tr>
<tr>
<td>Pt-10Al-4Ti</td>
<td>L12</td>
<td>3.8921</td>
<td>3.8642</td>
</tr>
</tbody>
</table>

2.2 Mechanical properties of ternary alloys

Figure 6 shows the high-temperature compression strength of Pt-10Al-4Ru compared to Mar-M247 (a Ni-based superalloy) and the tensile strength of PM2000 (an Fe-based superalloy) (22). It can be seen that the Pt-10Al-4Ru alloy based on γ/γ' precipitation strengthening has higher compressive strength and the ability of withstand higher temperatures than the traditional Ni-based and Fe-based superalloys at 1200°C.
Fig. 6. High-temperature compression strength of Pt-10Al-4Ru alloy compared to Mar-M247 (a Ni-based superalloy) and the tensile strength of PM2000 (an Fe-based superalloy) (22)

Süss (59) studied the stress-rupture strength and high-temperature creep properties of the Pt-Al-X (X = Cr, Ru, Ir) ternary alloy system. Figure 7 shows the stress-rupture curves of Pt-10Al-4Ru and Pt-10Al-4Cr alloys at 1300°C. The alloys’ interpolated strength levels for a rupture time of 10 h are summarized in Table 4. PM2000 shows the highest high temperature rupture strength among all the tested alloys, but the lower slope of the stress-rupture curve indicates that the alloy has high stress sensitivity and brittle creep behavior. This means PM2000 alloy is likely to be damaged by stress concentration or short-term overload in practical use. In contrast, the stress-rupture curve of Pt-based alloys has a steeper slope. Pt-10Al-4Cr alloy has the highest strength at 1300°C. The high-temperature durability of this precipitation strengthened alloy at 1300°C is higher than the ODS and DPH alloys (53) and the solid solution-strengthened Pt-20wt.% Rh alloys, and is close to the strength of Pt-30wt.% Rh alloys (10). However, due to wide price fluctuations of Rh as well as processing difficulties, the practical application of Pt-30wt.%Rh alloy is restricted (10).

Figure 8 shows the creep curves of the above alloys under the test conditions of 1300°C and 30 MPa (59). The initial stage of creep was not observed in the three Pt-based alloys. After the second-stage creep, the Pt-based alloys undergo the third stage of creep and subsequent rupture. The creep rupture strain was as high as 10%~30%. For PM2000 alloy, it is impossible to divide the creep curve into different stages because of the very low creep rate and fracture strain of less than 1%.

Fig. 7. Stress-rupture curves of PM2000 and Pt-10Al-4X (X = Cr, Ru, Ir) alloys at 1300°C in air (59)
Table 4 stress-rupture strength $R_m$ for PM2000 and Pt-10Al-4X alloys (59)

<table>
<thead>
<tr>
<th>Alloy</th>
<th>Stress-rupture strength $R_m/10^8/1300^\circ C$</th>
</tr>
</thead>
<tbody>
<tr>
<td>PM2000</td>
<td>25</td>
</tr>
<tr>
<td>Pt-10Al-4Cr</td>
<td>17</td>
</tr>
<tr>
<td>Pt-10Al-4Ru</td>
<td>15</td>
</tr>
<tr>
<td>Pt-10Al-4Ir</td>
<td>13</td>
</tr>
</tbody>
</table>

Fig. 8. Creep curves of PM2000 and Pt-10Al-4X (X = Cr, Ru, Ir) alloys tested at 1300ºC and 30 MPa (59).

2.3 Oxidation behavior of ternary alloy

The service environment for components such as aero engines, industrial gas turbine blades and aerospace engine thrusters is very harsh: high temperature, strong oxidation and corrosion. Although Pt-based superalloys with potential for high-temperature applications have been confirmed from the perspective of microstructure features and high-temperature mechanical properties, further attention must be given to assessing their oxidation and corrosion behavior (10). Hill et al. (46) conducted oxidation tests on Pt-Al-Ni, Pt-Al-Ru, Pt-Al-Re, Pt-Nb-Ru and Pt-Ti-Ru in flowing air at 900ºC, 1100ºC, 1300ºC and 1400ºC, respectively. The study found that all Al-containing alloys show negligible weight loss on oxidation, while the Pt-Ti-Ru and Pt-Nb-Ru alloy systems have significant mass increases, indicating lower resistance to oxidation. Figure 9 shows optical micrographs of the transverse-sectional morphology of several Pt-based alloys (46). It can be seen that the Pt-Ti-Ru alloy has grain boundary oxidation, while severe internal oxidation has occurred in the Pt-Nb-Ru and Pt-Ta-Re alloys. The Al-containing alloy has formed a protective aluminum oxide film on its outer surface. This layer prevents oxidation of the underlying metal, conferring better oxidation resistance on this alloy.

Figure 10 shows the isothermal oxidation curves of Al-containing Pt-based ternary alloys Pt-10Al-4X (X = Cr, Ir, Ru, Ti) and PM2000 Fe-based superalloys at 1350ºC (10). It can be seen that Pt-10Al-4Ti and Pt-10Al-4Ru exhibit a parabolic oxidation law similar to the PM2000 alloy. The Pt-10Al-4Ir and Pt-10Al-4Cr alloys exhibit a parabolic change in the initial stage of oxidation,
and the oxidation rate is relatively high. After that, the oxide film is grown at a logarithmic rate. After 800 h of oxidation a continuous oxide layer is obtained, which has better oxidation resistance than the PM2000 alloy. Experimental research on the microstructure, mechanical properties and oxidation resistance of the Pt-Al-X series of superalloys and comparison with the PM2000 benchmark alloy leads to the conclusion that the highest performing Pt-based ternary superalloys are Pt-10Al-4Cr and Pt-10Al-4Ru (10).

![Fig.9](image_url) Fig.9. Optical micrographs for comparison of the transverse sections of the oxidized samples (a) Pt-14Al-8Ru; (b) Pt-23Ti-7Ru; (c) Pt-24Ta-4Re; (d) Pt-24Nb-3Ru (46)

![Fig.10](image_url) Fig.10. Results of the isothermal oxidation tests conducted on Pt-Al-X (X = Cr, Ir, Ru, Ti) alloys at 1350ºC (10)

### 3 Pt-Al-Cr-X quaternary and multi-element alloys

Early studies on Pt-based superalloys mainly focused on the addition of alloying elements to improve oxidation resistance and ensure the γ' phase has a stable L12 structure. Research shows...
that the volume fraction of the precipitation phase reaches only about 30% no matter how the heat treatment process is optimized. Hence it is difficult to obtain the desired strength of the alloy (60). Subsequent work mainly focused on the Pt-Al-Cr ternary alloy system. Adding Ni, Ru and other alloying elements to form quaternary and multi-element Pt-based superalloys is expected to further improve the microstructure and to enhance the mechanical properties and oxidation resistance. Compared with steel, Ni-based alloys and Al alloys, the experimental data and phase diagrams of Pt-based alloys are relatively lacking. In order to develop quaternary and multi-element Pt-based alloys, relevant research institutions in South Africa, Germany, and the UK have collaborated to establish Pt-Al-Ru (61) and Pt-Cr-Ru (62) ternary system and Pt-Al-Cr-Ni (63) quaternary system alloy databases by experiment and first-principles thermodynamic calculations.

3.1 Pt-Al-Cr-Ni Quaternary alloy

Ni has a good solid solution strengthening effect on the Pt matrix (64) and its addition can stabilize the L1$_2$ structure of the Pt$_3$Al phase. Researchers have added Ni to the Pt-Al-Cr alloy to form a Pt-Al-Cr-Ni quaternary alloy (36). In order to reflect the results of previous research on Pt-10Al-4Cr, the nominal composition ratio of Pt:Al:Cr in the quaternary alloy is designed to be approximately 86:11:3, and the maximum content of Ni is 10 at.%. The Pt-11Al-3Cr-(0-10)Ni alloy system has a single-phase structure after solution treatment at 1450ºC. The microstructure obtained is similar to that of Ni-based superalloys after ageing treatment at 1000ºC, but alloys with Ni content below 6 at. % seem to have a lower coherence of precipitation. After ageing treatment, the Pt-11Al-3Cr-6Ni alloy has a maximum γ' phase volume fraction of about 23%. The cubic precipitate phase is arranged in a straight line with a side length of 200~500nm, and the degree of mismatch between the precipitate phase and the matrix is about -0.1 % (similar to Ni-based superalloys) (Figure 11). Spherical particles are observed in alloys with a Ni content of more than 6 at. % and it is believed that the change in the γ' phase morphology is due to the increase in the Ni concentration and a decrease in the degree of mismatch. Ageing at 1100ºC will cause coarsening of γ' phase and reduce the volume fraction of the γ' phase. However, as ageing temperature increases, the volume fraction of the γ' phase of the Ni-containing alloy decreases less than that of the Ni-free alloy.

A Pt-(12-15)Al-3Cr-(4-8)Ni alloy with Al content near the solubility limit (15 at. %) was selected to obtain a Pt-based alloy with high volume fraction of the γ' phase (36, 37). For alloys with Al content less than 13 at. % it is possible to obtain a homogenized single-phase structure after heat treatment at 1500ºC. Alloys with a higher Al content will form a eutectic two-phase dendritic structure even after heat treatment at 1530ºC. For alloys with Al content less than 13 at. %, ageing treatment at 120 h at 1000ºC ageing produces a uniformly distributed precipitation of Pt$_3$Al. Figure 12 shows cubic Pt$_3$Al particles with an average side length of 520 nm in the Pt-14Al-3Cr-6Ni alloy. The absolute mismatch between γ and γ' phases decreases with the increase of Ni content. For alloys with Ni content higher than 5 at. %, the slightly negative mismatch value (less than -0.5%) at room temperature and the cubic or spherical particle morphology of
the γ' phase indicate that the γ and γ' phases are in a coherent state. Increasing the main γ' phase forming element (Al) to 13 at. % can increase the volume fraction of γ' up to 30% (37).

The alloy system Pt-Al-Cr-Ni was studied with an Al content limited to 12.5 at. % and a Cr content up to 6 at.%. The effect of Cr content and heat treatment on the volume fraction of the γ' reinforcing phase is reported (34). Analysis of the microstructure showed that the dendritic cast structure of Pt-12.5Al-3Cr-6Ni, Pt-12Al-6Ni and Pt-12Al-6Cr-6Ni can be homogenized by heat treatment at 1500~1510ºC. After homogenization treatment (12h at 1500ºC), the Pt3Al precipitation in the Pt-12.5Al-3Cr-6Ni alloy was almost completely suppressed after water quenching (Figure 13a). Air cooling causes the Pt3Al particles (average size 200 nm) to be uniformly distributed with a volume fraction of about 30% (Figure 13b). Furnace cooling from 1500ºC resulted in cubic and coarse particles distributed in the alloy with a volume fraction of 34% (Figure 13c). Increasing the Cr content to 6 at. % resulted in Pt3Al with an average particle size of 500 nm and a volume fraction that reached 50% after solution heat treatment for 6 h at 1500ºC-1510ºC and air cooling.

Fig.11. Scanning electron (SE) image of Pt-11Al-3Cr-6Ni after solution heat treatment at 1450ºC for 24h and ageing at 1000ºC for 120h (36)

Fig.12. SE image of Pt-14Al-3Cr-6Ni after solution heat treatment at 1500ºC for 12h and ageing at 1000ºC for 120h (37)
The γ' phase dissolves at very high temperatures, therefore the designed Pt-based superalloys have a maximum operating temperature of 1300°C. Figure 14 shows the stress-rupture strength curves of Pt-10wt.%Rh, Pt-10wt.%Rh DPH and Pt-12Al-6Cr-5Ni alloys at 1300°C (65). The latter has the highest fracture strength. The minimum creep rate of Pt-12Al-6Cr-5Ni alloy is almost 3 orders of magnitude lower than that of the initial Pt-10Al-4Cr alloy at 1300°C. Under the stress of 30 MPa, the creep performance of Pt-12Al-6Cr-5Ni alloy is better than the PM2000 benchmark alloy (Figure 15). Adding a small amount of B (0.3 at. % or 0.7 at. %) can significantly improve the creep strength and ductility of the Pt-12Al-6Cr-5Ni alloy (66). To investigate the oxidation resistance of Pt-12Al-6Cr-5Ni, Wenderoth et al. (67) studied its isothermal oxidation behavior of after 400 h exposure to a temperature range of 1100~1300°C in air. A layer of Al2O3 was observed on the surface below which a free γ'-free layer was detected. It was also observed that the size of the γ'-free layer continuously increases with time and temperature. Furthermore, the local concentration of Al in the γ'-free layer increases with elevated ageing temperatures. After 400 h exposure at 1300°C a thick polycrystalline Al2O3 scale with large oxide grains developed on the surface. This is in good agreement with the typical behavior of alloy systems forming protective Al2O3 scales.
Fig. 15. Minimal tension creep rates at 1300°C of Pt-12Al-6Cr-5Ni, Pt-10Al-4Cr and PM2000 alloys (65)

3.2 Pt-Al-Cr-Ru Quaternary alloy

The strong solid solution strengthening element Ru can be added to the Pt-Al-Cr ternary alloy to form a Pt-Al-Cr-Ru quaternary alloy. This increases the volume fraction of the γ' phase and further improves the mechanical performance and oxidation resistance of the alloy (68). Figure 16 shows the typical TEM two-phase microstructure of Pt-12Al-4Cr-2Ru (at. %) alloy with the Pt₃Al precipitation phase and the (Pt) matrix (69). The volume fraction of the Pt₃Al precipitate has increased significantly, with the highest volume fraction of the precipitation phase in Pt-based alloys found so far. The morphology of the precipitation phase is mainly cubic with a side length of about 200 nm and a small amount being irregular. X-ray diffraction analysis confirmed that the Pt₃Al precipitation phase is cubic.

Fig. 16. TEM micrograph for Pt-12Al-4Cr-2Ru, showing Pt₃Al precipitates, an example is marked A (69)

The quaternary alloy was prepared by arc melting and then aged in air for 100 h at 1250°C
in a muffle furnace. It was then water quenched. Test samples for investigation of mechanical properties were obtained from the bulk material by machining. Table 5 lists the mechanical properties of several Pt-Al alloys, Fe-based (PM2000) and Ni-based (CMSX-4) superalloys (68). From the performance comparison of the three Pt-based alloys, it can be found that the quaternary alloy has the highest hardness value, but its room temperature tensile strength and elongation are the lowest. This is inconsistent with the expected result of Ru as a strong solid solution strengthening element. The tensile fracture morphology of the alloy was analyzed. It was found that only the Pt-11Al-3Cr-2Ru alloy showed intergranular fracture, while all the ternary alloys had cleavage fracture with some localized signs of dimpling. It is likely that the lower ultimate tensile strength value of the quaternary alloy is related to the intergranular failure mode, which also correlates to the lower plasticity expressed by the lower elongation value. Compared with Fe-based and Ni-based superalloys, the tensile strength of Pt-based superalloys is within the ultimate tensile strength range of high-temperature alloys at room temperature.

Odusote (70, 71) studied the isothermal oxidation behavior of Pt-11Al-3Cr-2Ru (at. %) at 1350ºC in air. It is found that the oxide layer of the alloy is mainly composed of α-Al₂O₃ and the thickness of the oxide layer increases with the oxidation time according to a parabolic law (similar to Pt-Al-Cr or Pt-Al-Ru alloys). The growth mechanism of the oxide layer is mainly the diffusion of oxygen atoms into the inner layer along the oxide grain boundary, accompanied by the outward diffusion of a small amount of Al atoms. The oxide protective layer is dense, has good adhesion to the substrate and no local discontinuities or detrimental internal oxidation phenomena have been found (Figure 17)(71). These characteristics of the oxide layer indicate that the Pt-11Al-3Cr-2Ru alloy has good oxidation resistance and has the potential for high-temperature applications.

Table 5 Mechanical properties for platinum-based alloys and selected high-temperature alloys (68)

<table>
<thead>
<tr>
<th>Alloy</th>
<th>Hardness, Ultimate Tensile Strength at room temperature</th>
<th>Elongation</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>HV</td>
<td>MPa</td>
</tr>
<tr>
<td>Pt-10Al-4Cr</td>
<td>317±13</td>
<td>836</td>
</tr>
<tr>
<td>Pt-10Al-4Ru</td>
<td>278±14</td>
<td>814</td>
</tr>
<tr>
<td>Pt-11Al-3Cr-2Ru</td>
<td>361±10</td>
<td>722</td>
</tr>
<tr>
<td>PM2000</td>
<td>–</td>
<td>700</td>
</tr>
<tr>
<td>CMSX-4</td>
<td>–</td>
<td>870</td>
</tr>
</tbody>
</table>
Alloys belonging to the Pt-Al-Cr-Ru series are currently being developed since they show the best overall performance. However Pt-based alloys have the disadvantages of high price and high density (10). The addition of a cheap and low-density alloy element to replace part of the Pt while maintaining the high-temperature properties and desired microstructure is a subject of current research. Wenderoth et al. (72) added the refractory metal Nb to the Pt-based alloy to improve the high-temperature strength through precipitation hardening. Vanadium belongs to the same group of the Periodic Table as Nb but has a smaller atomic radius and higher solid solubility in the Pt matrix. As well as its effect on precipitation hardening, V may also have a solid solution hardening effect. Odera et al. (73) prepared and analyzed four Pt-Al-Cr-Ru-V five-component alloys and two Pt-Al-Cr-Ru-V-Nb six-component alloys by adding V and Nb on the basis of a Pt-12Al-4Cr-2Ru quaternary alloy with excellent properties. V was added in the range 5.2~19.0 at. %, Nb was added in smaller amounts. The content of Pt was reduced to the range 63.2~78.7 at.%. The study found that the expected Pt3Al precipitate appeared and that a two-phase structure (matrix with precipitates) was obtained in the four cast alloys, while two alloys had a single-phase structure with V preferentially distributed in the Pt matrix. The hardness of multi-element Pt-based alloys is higher than the quaternary alloys. The optimal V content is about 15 at. % to obtain the desired microstructure. A higher V content leads to the formation of a brittle Pt-V mesophase. Nb loss was too high for its alloying effect to be determined. According to the preliminary experimental results on microstructure and hardness of the multi-element systems based on Pt alloys, these materials have promise for high-temperature applications although further optimization of their composition is necessary.

4 Research on Hot Corrosion Performance of Platinum-based Superalloys

For materials used in high-temperature environments, the focus is usually on high-temperature strength (74). However, the creep, oxidation and corrosion resistance of alloys are also important (75). Increasing operating temperature will lead to continuous corrosion. Therefore, it is necessary to evaluate the corrosion resistance of high-temperature materials during the selection process (1, 76). For Ni-based superalloys, the high-temperature
strength of the alloy is improved by increasing the content of Al and reducing the content of Cr, but the alloy is more sensitive to high-temperature corrosion, and it is necessary to develop and introduce protective coatings (77).

Since Pt-based alloys show excellent performance in various high-temperature applications such as glass manufacturing and corrosive substance processing, Pt-based alloys can be used to solve problems encountered in the aerospace industry (29, 78). Pt-based superalloys are relatively new high-temperature materials, and there are very few literature reports on their corrosion properties. Fuel or intake air usually contains impurity elements such as Na, S, and V, which may form molten salt corrosion products such as Na\(_2\)SO\(_4\), NaCl, and V\(_2\)O\(_5\), which in turn may lead to high temperature hot corrosion (HTHC) of materials (22, 79). Hot corrosion caused by molten salt or corrosive gas accelerates the oxidation degradation of high temperature materials, adversely affects the mechanical properties of the alloy and shortens the service life of high temperature components. There are two types of hot corrosion in Ni-based superalloys: Type I and Type II. Type I hot corrosion is also known as HTHC and usually occurs in the temperature range 850~950ºC. Type II hot corrosion is also known as low temperature hot corrosion (LTHC) and generally occurs in the temperature range 650~800ºC (79). Type I or HTHC is the main corrosive process in aircraft gas turbine engines.

Maledi et al. (80) studied the accelerated corrosion behavior of five Pt-based superalloys in analytically pure anhydrous Na\(_2\)SO\(_4\) at 950ºC and compared them with NBSA with 1.25 \(\mu\)m thick Pt\(_2\)Al coating protection or no coating. The experimental results are listed in Table 6 (76). The corrosion kinetics of Ni-based and Pt-based superalloys for the first 50 h are shown in Figure 18 and Figure 19, respectively. Due to the protective oxide layer formed on the surface of the Pt-based alloy, the weight gain associated with corrosion is very small. On the other hand, uncoated NBSA form oxides in the initial stage of corrosion, leading to increased mass. After further exposure to the corrosive environment, non-protective oxides form and cause catastrophic corrosion damage. Although the coated NBSA has better corrosion resistance than the uncoated alloy, it still degrades prematurely compared to the Pt-based superalloy. Experiments show that in the molten Na\(_2\)SO\(_4\) environment, Pt-based alloys show superior corrosion resistance compared to both coated and uncoated NBSA. Figure 20 shows the protective layer with strong adhesion formed on the surface of the Pt-10Al-4Cr alloy. The surface protective layer morphology of the Pt-10Al-4Ru and Pt-11Al-3Cr-2Ru alloys is similar (80). Although the morphology of the surface oxide layer of Pt-based superalloys containing Cr or Ru appears porous, their resistance to hot corrosion is higher than that of the Pt-based superalloys containing Co. NBSA was subjected to penetration of S under the surface oxide layer, leading to formation of Cr and Ni sulfides. This alloy showed the worst resistance to sulfidation and hot corrosion. Resistance to sulfidation was also the subject of studies in the experimental work of Potgieter et al. (81). The results of the study are shown in Table 6. The tests were performed in a 0.2% SO\(_2\)-N\(_2\) mixed atmosphere and research conclusions were
similar to those obtained by Maledi et al. (80).

<table>
<thead>
<tr>
<th>Alloy name</th>
<th>Nominal composition, at.%</th>
<th>Cumulative weight gain during corrosion, mg/cm²</th>
</tr>
</thead>
<tbody>
<tr>
<td>RS-1</td>
<td>Pt-10Al-4Cr</td>
<td>0.00004</td>
</tr>
<tr>
<td>RS-2</td>
<td>Pt-10Al-4Ru</td>
<td>0.00008</td>
</tr>
<tr>
<td>RS-3</td>
<td>Pt-11Al-3Cr-2Ru</td>
<td>0.0001</td>
</tr>
<tr>
<td>P420</td>
<td>Pt-15Al-6Co</td>
<td>0.0001</td>
</tr>
<tr>
<td>P421</td>
<td>Pt-15Al-12Co</td>
<td>0.004</td>
</tr>
<tr>
<td>CMSX-4(uncoated)</td>
<td>Ni-6.5Cr-11Co-0.3Mo-1.7W-1.8Ta-11.3Al-0.9Ti</td>
<td>0.470</td>
</tr>
<tr>
<td>CMSX-4(coated)</td>
<td>Ni-6.5Cr-11Co-0.3Mo-1.7W-1.8Ta-11.3Al-0.9Ti</td>
<td>0.038</td>
</tr>
</tbody>
</table>

Fig.18. Corrosion kinetics of the coated CMSX-4 NBSAs during exposure to Na₂SO₄ at 950ºC for the first 50 hours (80)

Fig.19. Corrosion kinetics of five Pt-based alloys of various compositions for the first 50 hours (80)
Fig. 20. SEM secondary electron image showing the thin protective scale on the surface of the platinum-based superalloy RS-1(Pt-10Al-4Cr) (80)

5 Conclusions

More than 20 years of research and development have yielded interesting results for Pt-based superalloys. The results obtained so far are only the tip of the iceberg of a very interesting and topical subject characterized by excellent prospects for future use. Optimized design of alloy composition, microstructure characteristics, mechanical properties and oxidation corrosion behavior have been achieved. Among the systems studied so far, the Pt-Al-Cr-Ru alloy system has been selected and optimized for excellent performance. The ultimate goal is to develop Pt-based superalloys for application in industrial fields such as gas turbine engines. However, this is a competitive market that is difficult to penetrate with new materials. One possibility would be to exploit the possibility for Pt-based alloys to be used at temperatures 200°C higher than for NBSA (82). During the transition period, Pt-based superalloys could be used in other fields, such as castings, powder metallurgy products and coatings to accelerate their final use in gas turbine engines.

At present, the composition and structure design of Pt-based superalloys mainly follows the successful experience of NBSA development. The melting point of Pt is 316°C higher than that of Ni, but the difference in melting point between Pt-based superalloys and Ni-based superalloys is less than 150°C. The main reason for this discrepancy is the addition of low melting point alloying elements (such as Al) which reduces the melting point of the resulting alloy to about 1500°C. The advantage of Pt's high melting point has thus not yet been fully exploited. The development cycle of a new generation of superalloys is very long and the material cost is relatively high for precious metals. The optimization process for the design of new Pt-based superalloys could be accelerated with the help of material genome research concepts to further increase the alloy's temperature tolerance and reduce research and development costs.

Pt-based superalloys can still be considered a brand new alloy system when compared to, say, NBSA or stainless steels. The accumulation of fundamental data such as phase precipitation mechanisms and alloy properties is far from being complete. There has been little research on
the influence of manufacturing processes (for example, precision casting, directional solidification and single crystal preparation) on the formability and mechanical properties of the alloys. There has similarly been insufficient verification and assessment of performance under actual use in typical environmental conditions. To accelerate bringing these materials to market, the level of research and development on Pt-based superalloys needs to be improved urgently.

Acknowledgments

This work is supported by Basic Research Key Program of Yunnan (No.2019FA048) and The Major Science and Technology Program of Yunnan (Nos. 2019ZE001, 202002AB080001-1).

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Doi: 10.1595/205651321X16221908118376
