



# The machinability of nickel-based alloys: a review

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## Abstract

This paper presents a detailed review of the various types of nickel-based alloys available on a commercial basis and their development including alloying additions as well as processing techniques employed to achieve specific mechanical and/or chemical properties. Problems associated with the machining of nickel-based alloys as well as tool wear and the mechanisms responsible for tool failure are identified and discussed. The integrity of the machined surfaces and tool life are the most important considerations during machining. These and other factors governing the machinability of nickel-based alloys are mainly affected by notching of the cutting tool, primarily at the depth of cut region, as well as by flank wear and insert chipping/fracture; and by failure modes caused singly or jointly by diffusion, attrition, and abrasion wear mechanisms in addition to mechanical and thermal fatigue loading of the cutting tools. Most of the failure modes can be minimised when machining in the presence of coolants and in an oxygen-rich environment. The use of a high pressure coolant supply, despite improved chip segmentation and machining in the presence of argon and nitrogen-rich environments, tend to accelerate tool wear rate leading to lower tool life. Improvements in tool performance can be achieved with an increase in the included angle and/or the nose angle of cutting tools due to the increased edge strength and tool–chip contact area plus a reduction in the approach angle. Recently developed cutting tool materials such as mixed oxide, SiC whisker reinforced alumina ceramics, sialon and multi-layer coated cemented carbide cutting tool material have all exhibited the capability to machine nickel-based alloys at higher speed conditions than those achieved with conventional cemented carbide tools. © 1999 Elsevier Science S.A. All rights reserved.

**Keywords:** Nickel-based alloys; Processing techniques; Machinability

## 1. Introduction

Nickel-based alloys play an extremely important role in gas turbine engines. In addition to their use in aircraft, marine, industrial and vehicular gas turbines, nickel-based alloys are now also used in space vehicles, rocket engines, experimental aircraft, nuclear reactors, submarines, steam power plants, petrochemical equipment and other high-temperature applications. Nickel-based alloys are known as some of the most difficult-to-machine superalloys in order to satisfy production and quality requirement [1,2]. The properties of Ni-based alloys contributing to poor machinability may be summarised as follows [3–5]: (i) a major part of their strength is maintained during machining due to their high temperature properties; (ii) work hardening occurs rapidly during machining, which is a major factor

contributing to notch wear at the tool nose and/or depth of cut line; (iii) cutting tools suffer from high abrasive wear owing to the presence of hard abrasive carbides in the superalloy; (iv) chemical reaction occurs at high cutting temperatures when machining with commercially available cutting tool materials, leading to a high diffusion wear rate; (v) welding/adhesion of nickel alloys onto the cutting tool frequently occur during machining, causing severe notching as well as spalling on the tool rake face due to consequent pull-out of the tool materials; (vi) production of a tough and continuous chip, which is difficult to control during machining, thereby contributing to the degradation of the cutting tool by seizure and cratering; and (vii) the poor thermal diffusivity of nickel-based alloys often generates high temperature at the tool tip as well as high thermal gradients in the cutting tool.

Of factors impairing the machinability of nickel-based alloys, short tool life and severe surface abuse of machined workpiece are the most important consider-

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ations [1,2]. The surface alteration (such as residual stresses) induced during machining may result in distortion which adversely affect the mechanical properties and stress-corrosion properties of the machined component. Extreme care must be taken therefore, to ensure the surface integrity of the component during machining. Tool lives obtained when machining nickel-based alloys are severely inadequate. Most of the major parameters including the choice of tool materials, tool geometry, machining method, cutting speed, feed, depth of cut, etc. must be controlled in order to achieve adequate tool lives during machining.

With the increasing complexity and the higher operating temperatures of modern aircraft engines, superalloy development has evolved from the simple nickel–chromium matrix to the multi-element, multiphase systems of today to meet more stringent operating conditions. Nickel-based superalloys now contain additions of chromium, aluminium, titanium, cobalt, molybdenum and other elements in varying quantities to give higher performance. With a material of such composition, the problems of short tool life and metallurgical damage to the workpiece during machining are more prominent. Although nickel-based alloys are not exceptionally hard (the hardness range being about 250–350 HV), their outstanding high temperature strength and extreme toughness create difficulties during machining due to their work hardening tendency which results in very high cutting forces and significant burr formation during machining.

## 2. Development and application of nickel-based alloys

Nickel-based alloys constitute ~45–50% of the total material required in the manufacture of an aircraft engine due to their outstanding strength and oxidation resistance at elevated temperatures in excess of 550°C [3,6]. Nickel-based alloys are manufactured from wrought (bar, sheet, large forgings) and castings.

### 2.1. Wrought nickel-based alloys

The range of wrought nickel-based alloys in use today was developed from the basic 80% nickel, 20% chromium alloy used for over 50 years in electrical resistance applications. They were first employed as materials for turbine blades in the early 1940s due to their admirable oxidation resistance and creep rupture strength which were considerably improved by the addition of small amounts of titanium and aluminium [5]. The first alloy produced was Nimonic 80 which was subsequently modified to Nimonic 80A. Later it was established that the addition of ~20% cobalt resulted in an alloy (Nimonic 90) with ~50°C temperature advantage over Nimonic 80A. With increasing demands

from the turbine engine designer for improvement in high temperature capability, increased amounts of titanium and aluminium were added and further solid solution strengthening was achieved by the addition of molybdenum. This led to the development of Nimonic 105, 115, etc. Waspaloy and M-252 are also two important wrought nickel-based alloys, which were developed, respectively, by Pratt and Whitney Aircraft and the General Electric Company in the late 1940s [3]. Unique to these alloys was the addition of molybdenum as a solid-solution and carbide-forming strengthener. These wrought alloys were predominantly used in turbine blades.

Rene' 41 and 95 are high-strength precipitation-hardened sheet alloys, often used in welded fabrications. Inconel 718, an iron-containing alloy, provides moderate high temperature strength as well as good resistance to strain-age cracking in welding. Incoloy 901 is another prominent member of nickel–iron based alloys. These alloys, as well as some wrought nickel alloys such as Waspaloy and Astroloy, find their largest application in turbine disks because they can provide the necessary high strength at turbine-disk operating temperatures.

The progressive improvement of nickel-based alloys has not simply been a function of composition, it is also associated with the method of melting, the hot working conditions applied and the optimisation of the heat treatment process(es). The early nickel-based alloys such as Nimonic 80A were air melted using the high-frequency melting technique. It was well known even in those early days that further strengthening could be achieved by increasing the hardening elements but this was not possible with air melting. A major breakthrough in the development of nickel-based alloys occurred in the 1950s with the introduction of vacuum induction melting and refining. Vacuum melting prevents the oxidation of the reactive hardening elements. It thus became possible to incorporate more titanium and aluminium (for  $\gamma'$  precipitation) in the production process. Vacuum refining also removes some of the volatile trace elements (bismuth, lead, tellurium, etc.) present in the raw materials used that adversely affected the creep strength and ductility of nickel-based alloys. A further benefit of reducing the trace element content is a marked improvement in hot workability. Thus, the presence of vacuum melting and refining opened the way for the development of still stronger wrought nickel-based alloys, culminating in compositions such as Nimonic 115 and Udimet 700 [5]. Typical wrought nickel-based alloys are listed in Table 1 [3].

### 2.2. Cast nickel-based alloys

By the late 1950s, turbine-blade operating temperatures, which were always challenging, were clearly limited by the capability of the best available wrought

Table 1  
Nominal chemical composition (weight %) and density of wrought nickel-based alloys

Alloy designation	Ni	Cr	Co	Mo	W	Ta	Cb	Al	Ti	Fe	Mn	Si	C	B	Zr	Other	Density (g cm <sup>-3</sup> )
Astroloy	55.1	15.0	17.0	5.25	—	—	—	4.0	3.5	—	—	—	0.06	0.030	—	—	7.91
D-979	45.0	15.0	—	4.0	4.0	—	—	1.0	3.0	27.0	—	—	0.05	0.010	—	—	8.19
Hastelloy alloy X	47.3	22.0	1.5	9.0	0.6	—	—	—	—	18.0	0.50	0.50	0.10	—	—	—	8.22
Inconel alloy 600	76.6	15.8	—	—	—	—	—	—	—	7.2	0.20	0.20	0.04	—	—	—	8.33
Inconel alloy 601	60.7	23.0	—	—	—	—	—	1.35	—	14.1	0.50	0.25	0.05	—	—	—	8.04
Inconel alloy 625	61.6	22.0	—	9.0	—	—	4.0	0.2	0.2	3.0	0.15	0.30	0.05	—	—	—	8.44
Inconel alloy 706	41.5	16.0	0.5	0.5	—	—	2.9	0.2	1.75	40.0	0.18	0.18	0.03	—	—	—	8.08
Inconel alloy 718	53.0	18.6	—	3.1	—	—	5.0	0.4	0.9	18.5	0.20	0.30	0.04	—	—	—	8.19
Inconel alloy X-750	73.0	15.0	—	—	—	—	0.9	0.8	2.5	6.8	0.70	0.30	0.04	—	—	—	8.30
IN-102	67.9	15.0	—	3.0	3.0	—	3.0	0.4	0.6	7.0	—	—	0.06	0.005	0.03	0.02 Mg	8.55
IN-587	47.2	28.5	20.0	—	—	—	0.7	1.2	2.3	—	—	—	0.05	0.003	0.05	—	8.08
IN-597	48.4	24.5	20.0	1.5	—	—	1.0	1.5	3.0	—	—	—	0.05	0.012	0.05	0.02 Mg	8.04
IN-853	74.6	20.0	—	—	—	—	—	1.5	2.5	—	—	—	0.05	0.007	0.07	1.3 Y <sub>2</sub> O <sub>3</sub>	8.09
M-252	55.2	20.0	10.0	10.0	—	—	—	1.0	2.6	—	0.50	0.50	0.15	0.005	—	—	8.25
Nimonic alloy 75	78.8	20.0	—	—	—	—	—	—	0.4	—	0.10	0.70	0.01	—	—	—	8.37
Nimonic alloy 80A	74.7	19.5	1.1	—	—	—	—	1.3	2.5	—	0.10	0.70	0.06	—	—	—	8.22
Nimonic alloy 90	57.4	19.5	18.0	—	—	—	—	1.4	2.4	—	0.50	0.70	0.07	—	—	—	8.18
Nimonic alloy 105	53.5	14.5	20.0	5.0	—	—	—	1.2	4.5	—	0.50	0.70	0.20	—	—	—	7.99
Nimonic alloy 115	57.3	15.0	15.0	3.5	—	—	—	5.0	4.0	—	—	—	0.15	—	—	—	7.85
Nimonic alloy PE.11	39.0	18.0	1.0	5.25	—	—	—	0.85	2.35	33.5	—	—	0.05	—	—	—	8.02
Nimonic alloy PE.16	43.5	16.5	1.0	3.3	—	—	—	1.2	1.2	33.0	0.10	0.15	0.05	0.020	—	—	8.02
Nimonic alloy PK.33	55.9	18.5	14.0	7.0	—	—	—	2.0	2.0	0.25	0.10	0.15	0.05	0.030	—	—	8.21
Nimonic alloy 120	63.8	12.5	10.0	5.7	—	—	—	4.5	3.5	—	—	—	0.04	—	—	—	7.95
Nimonic alloy 942	49.5	12.5	1.0	6.0	—	—	—	0.6	3.9	27.5	—	—	0.03	—	—	—	8.12
Pyromet 860	43.0	12.6	4.0	6.0	—	—	—	1.25	3.0	30.0	0.05	0.05	0.05	0.010	—	—	8.23
RA-333	45.0	25.5	3.0	3.0	3.0	—	—	—	—	18.0	1.50	1.20	0.05	—	—	—	8.25
Rene' 41	55.3	19.0	11.0	10.0	—	—	—	1.5	3.1	—	—	—	0.09	0.005	—	—	8.25
Rene' 95	61.3	14.0	8.0	3.5	3.5	3.5	—	3.5	2.5	—	—	—	0.15	0.010	0.05	—	8.23
TD nickel	98.0	—	—	—	—	—	—	—	—	—	—	—	—	—	—	2.0 ThO <sub>2</sub>	8.90
TD nic	78.0	20.0	—	—	—	—	—	—	—	—	—	—	—	—	—	2.0 ThO <sub>2</sub>	8.41
Unimet 500	53.6	18.0	18.5	4.0	—	—	—	2.9	5.9	—	—	—	0.08	0.006	0.05	—	8.02
Unimet 520	56.9	19.0	12.0	6.0	1.0	—	—	2.0	3.0	—	—	—	0.05	0.005	—	—	8.22
Unimet 700	53.4	15.0	18.5	5.2	—	—	—	4.3	3.5	—	—	—	0.08	0.030	—	—	7.91
Unimet 710	54.9	18.0	15.0	3.0	1.5	—	—	2.5	5.0	—	—	—	0.07	0.020	—	—	8.08
Unitemp AF2-1DA	59.5	12.0	10.0	3.0	6.0	1.5	—	4.6	3.0	—	—	—	0.32	0.015	0.10	—	7.99
Waspaloy	58.3	19.5	13.5	4.3	—	—	—	1.3	3.0	—	—	—	0.08	0.006	0.06	—	8.19

alloys which were forged with great difficulty. Alloys with improved strength were available, but were impossible to forge by existing methods. The required strength was obtained by turning to investment cast alloys [3]. Cast nickel-based alloys did not gain widespread acceptance until the development of vacuum melting and casting techniques in the late 1950s. One of the first alloys to be vacuum cast was Inco 713C. The high carbon content of the alloy gave rise to excessive primary carbides and poor low temperature ductility, and caused problems in integrally bladed disk castings. A low carbon version of the alloy was developed to overcome this problem and is also used widely for turbine blades.

As with the wrought alloys, the demand for stronger cast alloys was initially met by increasing the titanium and aluminium content and the development of In 100. In the mid 1960s, General Electric introduced Rene' 77 and later Rene' 80 with similar titanium and aluminium content by the end of the 1960s. This new alloy exhibited higher rupture strength and corrosion resistance than Rene' 77 [5].

Since it was impractical to further increase the titanium and aluminium content in order to increase strength, another approach had to be adopted during the late 1960s. It was then established that a reduction in the titanium content improved castability. This led to cast alloys such as B-1900, with a low titanium-to-aluminium ratio and with greater additions of molybdenum and tantalum to increase the solid solution strengthening. Martin Metals favoured the replacement of molybdenum with tungsten and introduced Mar-M 200. This was found to give problems with inadequate intermediate temperature creep rupture ductility and creep life consistency [3]. In 1969, Martin Metals patented a method by which these properties were improved. This development consisted of the addition of  $\sim 2\%$  hafnium to the cast alloy composition. The beneficial effect of hafnium was found to apply to a whole range of cast alloys, e.g. Inco 713LC and B1900. IN-792 and Mar-M 432 combine the strength of IN-100 with the excellent hot-corrosion resistance of Udimet 500 (an early cast alloy) by maintaining a moderately high chromium content together with a relatively high titanium–aluminium ratio and using a careful balance of refractory metal additions. Typical cast nickel-based and wrought nickel–iron based alloys and their main compositions are listed in Table 2 [3].

### 2.3. Alloying elements and microstructure of nickel-based alloys

It is the continuous study of alloying elements and the microstructure of nickel-based alloys that has led to the development of nickel alloys. From general consideration, it can be stated that nickel-based alloys are

characterised, and strengthened by some alloying elements which tend to improved their microstructure or phase described below:

1. Alloy matrix ( $\gamma$ ). The continuous matrix is a face centred cubic (FCC) nickel-based austenitic phase, which usually contains a high percentage of solid-solution elements such as cobalt, chromium, molybdenum and tungsten.
2. Gamma prime ( $\gamma'$ ). Aluminium and titanium are added in amounts and mutual proportions to precipitate high volume fractions of FCC  $\gamma'$ , which invariably precipitates coherently with the austenite.
3. Carbides. Carbon is added at levels of  $\sim 0.05$ – $0.2\%$ , and is often made to react with reactive and refractory elements present to form primary MC carbides. During heat treatment and in service, these decompose and generate lower carbides such as  $M_{23}C_6$  and  $M_6C$ , which populate the grain boundaries.
4. Grain boundary  $\gamma'$ . For many alloys, heat treatment techniques have been developed to generate a film of  $\gamma'$  along the grain boundaries which is believed to optimise rupture properties.
5. Topologically close packed (TCP) phases. Under certain conditions, plate-like phases such as  $\sigma$ ,  $\mu$  and Laves form and promote lower rupture strength and ductility.

#### 2.3.1. Alloying elements in nickel-based alloys

Nickel alloys have little inherent resistance to high temperature oxidation. The addition of chromium produces an adherent protective scale containing  $Cr_2O_3$  on the surface of the material when heated in an oxidising environment. Optimum oxidation resistance in nickel–chromium alloys is obtained with chromium content in the range 15 to 30% [3]. Chromium also forms chromium carbides which can play an important role in strengthening the material at high temperature.

Aluminium, titanium and niobium are added to strengthen the material by the formation of  $\gamma'$  gamma prime, ( $Ni_3(Al,Ti)$ ). Niobium can be substituted to some extent for titanium or aluminium in  $\gamma'$  or if the niobium content is high enough (above  $\sim 4\%$ ) it can form a separate hardening phase,  $Ni_3Nb$ . Titanium and niobium can also form carbides. Aluminium has a beneficial effect on oxidation resistance as it forms stable  $Al_2O_3$  in the oxide scale at high temperatures. The formation of either  $Al_2O_3$  or  $Cr_2O_3$  in the scale depends on the relative amounts of the two elements in the alloy. During the development of high temperature alloys it was considered necessary to lower the chromium content in order to maintain microstructural stability and to allow more  $\gamma'$ -elements, i.e. aluminium, titanium and niobium, to be added to obtain creep rupture properties. The loss in oxidation resistance due to lower chromium content has been compensated for

Table 2  
Nominal chemical composition (weight %) and density of cast nickel-based and wrought nickel–iron based alloys

Alloy designation	Ni	Cr	Co	Mo	W	Ta	Cb	Al	Ti	Fe	Mn	Si	C	B	Zr	Other	Density (g cm <sup>-2</sup> )
<b>Nickel-based</b>																	
Alloy 713C	74	12.5	—	4.2	—	—	2.0	6.1	0.8	—	—	—	0.12	0.012	0.10	—	7.91
Alloy 713LC	75	12.0	—	4.5	—	—	2.0	5.9	0.6	—	—	—	0.05	0.010	0.10	—	8.00
B-1900	64	8.0	10.0	6.0	—	4.0	—	6.0	1.0	—	—	—	0.10	0.015	0.10	—	8.22
FORD 406	60	6.0	10.0	1.0	8.5	6.0	2.0	4.5	2.0	—	—	—	0.13	0.018	0.06	—	7.75
IN-100	60	10	15.0	3.0	—	—	—	5.5	4.7	—	—	—	0.18	0.014	0.06	1.0 V	8.08
IN-162	73	10.0	—	4.0	2.0	2.0	1.0	6.5	1.0	—	—	—	0.12	0.020	0.10	—	7.75
IN-731	67	9.5	10.0	2.5	—	—	—	5.5	4.6	—	—	—	0.18	0.015	0.06	1.0 V	8.10
IN-738	61	16.0	8.5	1.7	2.6	1.7	0.9	3.4	3.4	—	—	—	0.17	0.010	0.10	—	8.25
IN-792	61	12.4	9.0	1.9	3.8	3.9	—	3.1	4.5	—	—	—	0.12	0.020	0.10	—	8.63
M22	71	5.7	—	2.0	11.0	3.0	—	6.3	—	—	—	—	0.13	—	0.60	—	8.53
MAR-M 200	60	9.0	10.0	—	12.0	—	1.0	5.0	2.0	—	—	—	0.15	0.015	0.05	—	8.53
MAR-M 200(DS)	60	9.0	10.0	—	12.0	—	1.0	5.0	2.0	—	—	—	0.13	0.015	0.05	—	8.44
MAR-M 246	60	9.0	10.0	2.5	10.0	1.5	—	5.5	1.5	—	—	—	0.15	0.015	0.05	—	8.08
MAR-M 421	61	15.8	9.5	2.0	3.8	—	2.0	4.3	1.8	—	—	—	0.15	0.015	0.05	—	8.16
MAR-M 432	50	15.5	20.0	—	3.0	2.0	2.0	2.8	4.3	—	—	—	0.15	0.015	0.05	—	8.19
NX188(DS)	74	—	—	18.0	—	—	—	8.0	—	—	—	—	0.04	—	—	—	7.91
Rene' 77	58	14.6	15.0	4.2	—	—	—	4.3	3.3	—	—	—	0.07	0.016	0.04	—	8.16
Rene' 80	60	14.0	9.5	4.0	4.0	—	—	3.0	5.0	—	—	—	0.17	0.015	0.03	—	7.75
Rene' 100	60	9.5	15.0	3.0	—	—	—	5.5	4.2	—	—	—	0.18	0.014	0.06	1.0 V	8.02
SEL	51	15.0	22.0	4.5	—	—	—	4.4	2.4	—	—	—	0.08	0.015	—	—	8.00
SEL-15	58	11.0	14.5	6.5	1.5	—	0.5	5.4	2.5	—	—	—	0.07	0.015	—	—	8.63
TAZ-8A	68	6.0	—	4.0	4.0	8.0	2.5	6.0	—	—	—	—	0.12	0.004	1.00	—	8.66
TRW-NASA VIA	61	6.1	7.5	2.0	5.8	9.0	0.5	5.4	1.0	—	—	—	0.13	0.020	0.13	0.5 Re, 0.4 Hf	8.02
UDIMET 500	52	18.0	19.0	4.2	—	—	—	3.0	3.0	—	—	—	0.07	0.007	0.05	—	9.02
WAZ-20(DS)	72	—	—	—	20.0	—	—	6.5	—	—	—	—	0.20	—	1.50	—	8.22
<b>Nickel–iron based</b>																	
Incoloy alloy 901	42.5	12.5	—	5.7	—	—	—	0.2	2.8	36.0	0.10	0.10	0.05	0.015	—	—	7.91
A-286	26.0	15.0	—	1.3	—	—	—	0.2	2.0	53.6	1.35	0.50	0.05	0.015	—	—	7.97
Discaloy	26.0	13.5	—	2.7	—	—	—	0.1	1.7	54.3	0.90	0.80	0.04	0.005	—	—	8.19
N-155	20.0	21.0	20.0	3.0	2.5	—	1.0	—	—	30.3	1.50	0.50	0.15	—	—	0.5 V	7.94
V-57	27.0	14.8	—	1.25	—	—	—	0.25	3.0	52.0	0.35	0.75	0.08	0.010	—	—	7.94

Table 3

The relationship between composition and temperature capability of wrought nickel-based alloys

Alloys	AlTiNb	Mo W	Cr	Co	Other	Temperature capability for 1000 hrs life at 138 N/mm <sup>2</sup>				
						750	800	850	900	950
Nimonic 80A	3.8	--	19.5	--		████████				
Inconel X750	4.2	--	15.0	--	7Fe	████████	████████			
Nimonic 90	3.8	--	19.5	16.5		████████	████████	████████		
Udimet 500	6.0	4.2	19.0	18.0		████████	████████	████████	████████	
Nimonic 105	6.0	5.0	15.0	20.0		████████	████████	████████	████████	
Udimet 700	7.8	5.2	15.0	18.5		████████	████████	████████	████████	████████
Nimonic 115	8.8	3.3	14.3	13.3		████████	████████	████████	████████	████████

by increasing the aluminium content. Some superalloys such as Inconel 718 and Nimonic 901 contain high (18–36 wt.%) iron contents. Increasing the iron content in nickel-based alloys tends to decrease their oxidation resistance because of the less adherent oxide scale.

The addition of cobalt affords slight solid solution strengthening and reduces the solubility of aluminium and titanium in the nickel–chromium FCC austenitic matrix, thus increasing the  $\gamma'$  solvus temperature. This therefore help to maintain strength at elevated temperature. Cobalt also has greater solubility for carbon than nickel. This means that either more carbon can be tolerated or, for an equivalent carbon content, fewer carbides will be present.

Molybdenum, tungsten and tantalum are all used primarily to provide solid-solution strengthening at high temperature. They also form complex carbides, and tantalum in particular is soluble in  $\gamma'$ .

Boron and zirconium are added to improve creep strength and ductility. It is well established that these elements segregate in the grain boundaries and consequently modify the formation of the grain boundary carbides. The presence of boron and zirconium in nickel alloys is not entirely beneficial since they can adversely affect weldability. Carbon is added in limited amounts to improve creep strength due to the formation of carbides. Other trace elements that are added deliberately to improve workability, oxidation resistance, etc., are calcium, magnesium and yttrium. In recent years the addition of small amounts of hafnium, mainly to the cast alloys, has been found to be beneficial, providing improvements in both ductility and strength at low and intermediate temperatures, as well as raising the hot tear resistance in directional solidification.

The effect of alloying elements on temperature capability of wrought nickel-based alloys and cast nickel-based alloys are shown in Tables 3 and 4, respectively. Table 5 illustrates the relationship between composition and strength of some nickel-based alloys [5].

## 2.3.2. Microstructure of nickel-based alloys

### 2.3.2.1. Gamma prime ( $\gamma'$ ).

$\gamma'$  has an ordered FCC structure and is the precipitation hardening constituent in nickel-based alloys. The normal composition is  $\text{Ni}_3\text{Al}$  although it is usually designated  $\text{Ni}_3(\text{Al,Ti})$  since as much as 65% of the aluminium can be replaced by titanium. The nickel can be replaced to a limited degree by chromium, cobalt, molybdenum and iron [5].

During ageing after a solution treatment,  $\gamma'$  nucleates initially as a coherent precipitate with its crystallographic orientation identical to that of the  $\gamma$  matrix and develops as fine homogeneous particles with a spherical morphology which grow and become cuboidal as coherency is lost.  $\gamma'$  can also form at boundaries and second-phase particle interfaces during relatively high temperature ageing treatments or during the partial solution treatment applied to some alloys.

$\gamma'$  ( $\text{Ni}_3\text{Al}$ ) is a unique intermetallic phase. By dislocation interaction, it contributes antiphase boundary (APB) strengthening to the  $\gamma$ – $\gamma'$  alloy, and the strength of the  $\gamma'$  increases as the temperature increases. Furthermore, the inherent ductility of  $\gamma'$  helps to prevent severe embrittlement, unlike strengthening that is developed by phases of higher hardness, such as carbides. This is also in direct contrast to the severe embrittlement created by the formation of brittle  $\sigma$  and Laves phases [3].

Much of the aluminium in  $\text{Ni}_3\text{Al}$  can be replaced by titanium, niobium or tantalum leaving the possibility of producing metastable  $\gamma'$ . The titanium-rich metastable  $\gamma'$  can transform, on prolonged exposure to temperatures ranging from 700 to 900°C, to  $\text{Ni}_3\text{Ti}$  ( $\eta$ ) which has a hexagonal close packed (HCP) structure. The transformation involves a change from a coherent to a non-coherent precipitate. Increasing the titanium to aluminium ratio, therefore, decreases the stability of the alloy. Transformation of  $\gamma'$  to  $\eta$  results in a reduction in the alloy strength, which is undesirable.

Table 4

The relationship between composition and temperature capability of cast nickel-based alloys

Alloys	Al Ti Nb	Mo W	Cr	Co	Other	Temperature capability for 1000 hrs life at 138 N/mm <sup>2</sup>					
						750	800	850	900	950	100
Udimet 500	6.0	4.2	19.0	18.0	--						
Rene' 77	7.9	4.2	14.6	18.5	--						
Alloy 713LC	8.9	4.2	12.5	--	--						
IN-738	9.5	4.4	16.0	8.5	--						
B-1900	7.0	6.0	8.0	10.0	--						
IN-100	10.0	3.0	10.0	15.0	0.9V						
Mar-M 200	8.0	12.5	9.0	10.0	--						

When niobium content is  $>4\%$  in alloys such as Inconel 718, a similar transformation of  $\gamma'$  to  $\text{Ni}_3\text{Nb}$  is found. The transformation process is  $\gamma' \rightarrow \gamma'' \rightarrow \text{Ni}_3\text{Nb}$ .  $\gamma''$  has a body centred tetragonal (BCT) structure. This phase precipitates as small platelets that are coherent with the matrix. It can co-exist with  $\gamma'$  if the aluminium and/or titanium contents are also high enough. The  $\text{Ni}_3\text{Nb}$  has an orthorhombic structure. When present as large platelets it is detrimental to properties, but small discrete grain boundary  $\text{Ni}_3\text{Nb}$  particles are essential for satisfactory creep rupture properties.

**2.3.2.2. Carbides.** The role of carbides in superalloys is complex and dynamic. Most investigators believe that carbides do exert a significant and beneficial effect on rupture strength at elevated temperatures. In addition, it is quite clear that carbide morphology can influence ductility, and also that carbides can influence the chemical stability of the matrix through the removal of reacting elements. The three main types of carbides found in nickel-based alloy are MC,  $\text{M}_{23}\text{C}_6$  and  $\text{M}_6\text{C}$ .  $\text{Cr}_7\text{C}_3$  can also be present but is rare.

The MC carbides are a major source of carbon for the alloy to use later if it interacts chemically. They have a FCC structure and are normally considered amongst the most stable compounds in nature. They are formed either in the liquid state or at high temperature after solidification. They exist in blocky or script morphologies within the grains and along the grain boundaries. The elements titanium, tantalum, niobium and hafnium are the main MC forming elements with the less reactive elements such as molybdenum and tungsten also being involved.

The  $\text{M}_{23}\text{C}_6$  carbides are usually predominant in alloys of high chromium content, where they form during lower-temperature (760–980°C) heat treatment and service, both from the degeneration of MC carbides and from soluble carbon residue in the alloy matrix. They

have a significant effect on the properties of nickel alloys. When present as discrete particles in grain boundaries, their critical location promotes a significant effect on rupture strength, apparently through inhibition of grain boundary sliding. However, properties are adversely affected when  $\text{M}_{23}\text{C}_6$  carbides exist in the form of cellular structures or as films in grain boundaries.

The  $\text{M}_6\text{C}$  carbides can be formed during higher-temperature heat treatment in alloys containing relatively high molybdenum content. They can also be formed as a result of the breakdown of MC. Chromium, nickel and cobalt can substitute for the major M elements, Molybdenum and tungsten; unlike more rigid  $\text{M}_{23}\text{C}_6$ , the composition can vary widely. Since  $\text{M}_6\text{C}$  carbides are stable at higher temperatures than  $\text{M}_{23}\text{C}_6$  carbides,  $\text{M}_6\text{C}$  is more beneficial as a grain boundary precipitate to control grain size in the wrought alloy.

**2.3.2.3. Borides.** Many nickel-based high temperature alloys have small boron addition. Normally the addition of boron is associated with the  $\text{M}_{23}\text{C}_6$  type carbides. However, if present in amounts greater than  $\sim 0.01\%$ , a boride of the form  $\text{M}_3\text{B}_2$  can form. The M radical is usually molybdenum-rich but chromium, titanium and nickel may also be present. The  $\text{M}_3\text{B}_2$  boride has a tetragonal unit cell and tends to precipitate at the grain boundaries.

**2.3.2.4. Topologically close packed (TCP) phases.** The TCP phases which are formed in certain alloys as a result of exposure to elevated temperature and stresses are  $\sigma$ ,  $\gamma$ ,  $\mu$  and Laves. The  $\sigma$  phase can precipitate in the nickel–chromium based alloys in the temperature region between 650 and 925°C, especially when the alloy is subjected to stress. It is an intermetallic compound which forms from the transition metal elements and exhibits a tetragonal structure. It forms by a nucle-

Table 5  
The relationship between composition and strength of nickel-based alloys

Alloys	Al Ti Nb	Mo W	Cr	Co	Other	Yield strength (0.2%) Stress (N / mm <sup>2</sup> ) At 200°C (N / mm <sup>2</sup> ) for 0.1% creep strain in 100 hrs at 650 °C					
						700	900	1100	400	600	80
Waspaloy	4.4	4.5	19.5	13.5	--						
Inconel 718	6.75	3.0	19.0	--	Bal Fe						
Astroloy	7.5	5.0	15.0	17.0	--						
IN-100	9.3	3.2	12.5	18.5	0.8V						
Rene' 95	9.5	7.0	14.0	8.0	--						

ation and growth process as needles or plate-shaped particles at grain and twin boundaries as well as at the interfaces of carbides or borides.

The occurrence of  $\sigma$  and the other TCP phases has a detrimental effect on the mechanical properties of the alloy. The compounds themselves are brittle and hard and are therefore liable to fracture. The plate-like morphology is a source for crack initiation and propagation leading to reduced impact strength. Perhaps of even more significance is their effect on elevated temperature rupture strength. The TCP phases contain a high proportion of the refractory elements and therefore denude the matrix of these elements, causing weakening.

#### 2.4. Strengthening and heat treatment of nickel-based alloys

##### 2.4.1. Strengthening of nickel-based alloys

The strengthening mechanisms of nickel-based alloys include solid solution strengthening, precipitation hardening and dispersion hardening. The properties of the grain boundaries are largely controlled by the type, amount and morphology of the carbides and also by the presence of  $\gamma'$  in these regions.

**2.4.1.1. Solid solution strengthening.** As already discussed, commercial austenitic superalloys always contain substantial alloying additions in solid solution to provide strength, creep resistance or resistance to surface degradation. In addition, the stronger alloys contain elements, that, after suitable heat treatment or thermal-mechanical processing, result in the formation of small coherent particles of an intermetallic compound. Therefore, typical nickel-based alloys are variations of an austenitic nickel–chromium–tungsten (or molybdenum) matrix. In order to obtain useful solid solution strengthening, an alloying element should satisfy the following requirements: (i) it should have a wide range of solid solubility in the matrix; (ii) it should

have a large difference in atomic size with the matrix; and (iii) it should have a high melting point.

The solid solution elements typically found in the  $\gamma$  phase are likely to include aluminium, chromium, iron, titanium, tungsten, vanadium, cobalt and molybdenum [3]. The difference in atom diameter from that of nickel varies from +1% for cobalt to +13% for tungsten.

**2.4.1.2. Precipitation hardening.** Precipitation hardening in nickel-based alloys really implies  $\gamma'$  precipitation. Other phases such as borides and carbides provide little strengthening at low temperature because of their low volume fraction; they, however, influence the creep rate and rupture life of nickel alloys.

The increase in strength caused by the interaction between the dislocation and the  $\gamma'$  precipitate can be related to several basic factors such as: (i) the volume fraction of  $\gamma'$  present; (ii) the size of the  $\gamma'$  particle; and (iii) the anti-phase boundary (APB) energy.

At a given temperature, the volume fraction of  $\gamma'$  is directly proportional to the amounts of total hardening elements present, i.e. aluminium, titanium and niobium. The volume fraction directly affects the high temperature strength of nickel alloys. The influence of  $\gamma'$  volume fraction on the high temperature strength of nickel-based alloys is illustrated in Fig. 1 [3]. This effect is probably the major guide used by alloy designers.

The increase in the strength of alloys is also related to the size of the  $\gamma'$  particle. Normal heat treatment practice for commercial wrought nickel-based alloys is aimed at obtaining a  $\gamma'$  particle size corresponding to peak hardness or just below it. If, during service, the alloy is exposed to a temperature that is much higher than the prior ageing treatment, the  $\gamma'$  will coarsen and the strength will drop.

The anti-phase boundary (APB) energy is a function of the degree of ordering of the precipitate. When a dislocation cuts through a coherent ordered  $\gamma'$  particle it creates an area of high energy, impeding the move-



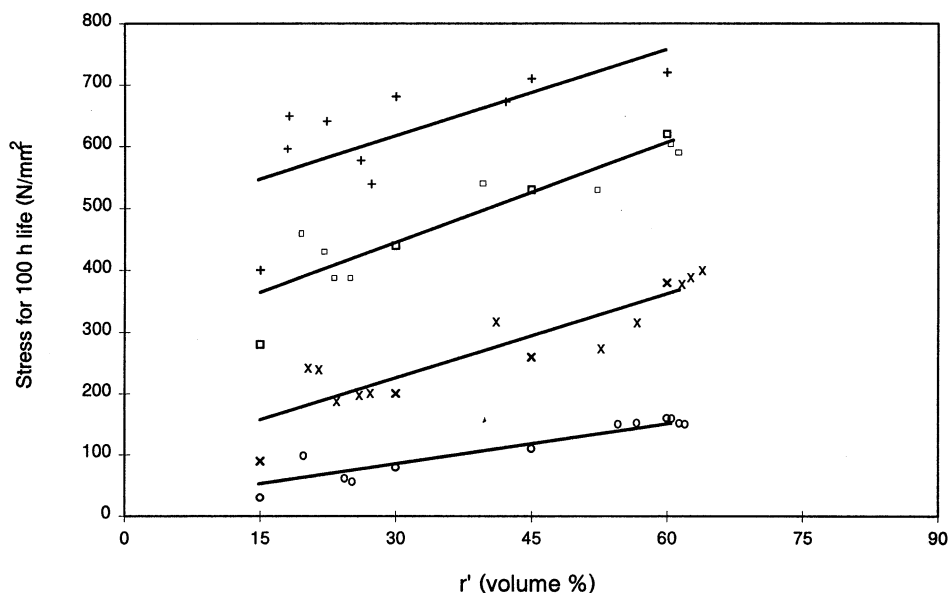


Fig. 1. The influence of the  $\gamma'$  volume fraction on the high-temperature strength of nickel-based alloys.

ment of the dislocation. The passage of a further dislocation along the same slip plane is assisted since it destroys the APB. The overall effect is a net addition to shear strength.

**2.4.1.3. Dispersion hardening.** The most effective way to maintain strength at high temperature above which  $\gamma'$  goes into solution is to employ a different, stable, non-metallic dispersoid phase that is essentially insoluble in the matrix. For high temperature service, the best dispersoid materials are oxides having high free energies of formation, such as thoria, yttria and lanthana. The size and distribution of the dispersoid phase govern its effectiveness as a strengthener. The particles must be small and uniformly distributed. The level of strengthening that can be achieved is limited by the low volume fraction of oxide that can be added while maintaining adequate ductility [5].

The development of the strength of nickel-based alloy strength can be summarised as follows: (i) solid-solution strengthening  $\gamma$ ; (ii) increase volume percent  $\gamma'$ ; (iii) increase fault energy of  $\gamma'$ ; (iv) solid solution hardening  $\gamma'$ ; (v) reduction in the formation of  $\eta$ ,  $\text{Ni}_3\text{Nb}$ , Laves and  $\sigma$  phases; (vi) control of carbide formation to prevent  $\text{M}_{23}\text{C}_6$  grain boundary films and cellular structures (for tensile strength); and (vii) control carbides and  $\gamma'$  (grain size, etc.) to enhance rupture strength.

#### 2.4.2. Heat treatment of nickel-based alloys

The heat treatments that contribute to improvement in the required properties of nickel-based alloys can be usefully discussed under three main headings.

**2.4.2.1. Annealing or softening.** Treatments of this type are only required for materials subjected to operations that work-harden the alloy. In this case, both intermediate and final treatments are given to produce a soft ductile end product. Softening is usually associated with recrystallisation but may also cause some solution of chromium carbides. Annealing is generally required for solid solution hardened alloys. Softening treatments are also applied during the fabrication of precipitation hardenable alloys and it is important to recognise that the temperature and time used for this operation may not be such that will fully dissolve the precipitation hardened phases [3].

**2.4.2.2. Solution treatment.** The main objective of solution treatment is to dissolve the precipitable phases, principally  $\gamma'$  and in some cases carbides, prior to their controlled precipitation during reheating/ageing or in special cases during cooling. The selection of a particular solution treatment temperature is dependent on the solvus temperature of the various phases present, the grain coarsening temperature. The finer the grain size, the better the low temperature tensile and fatigue properties; hence a low solution treatment temperature is preferred. The elevated temperature creep rupture properties on the other hand tend to be improved by the coarser grain produced by a higher solution treatment temperature.

**2.4.2.3. Ageing/precipitation hardening.** The objective of this treatment is to precipitate the hardening phases in a form that develops the required mechanical properties. A set of ageing curves for a nickel-based alloy—Nimonic 80A, illustrating the relationship between

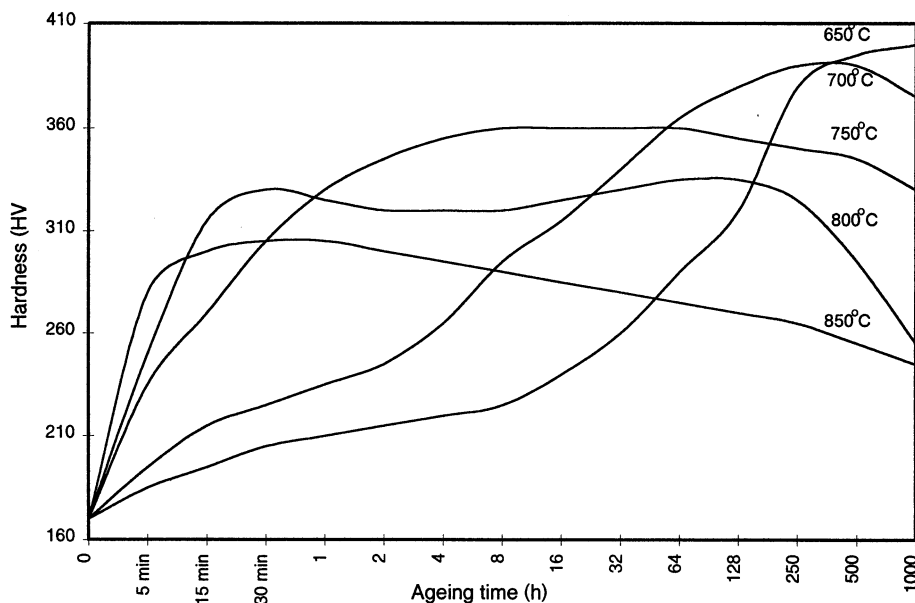


Fig. 2. The effect of ageing time and temperature on the hardness of Ni-based alloy (Nimonic 80A).

hardness (strength), ageing time and temperature, is shown in Fig. 2. It is clear that optimum hardening can be achieved after about 128 h at 700°C or 500 h at 650°C. These times are not acceptable commercially and the treatment adopted in practice for any alloy is generally a compromise between those giving optimum properties (not necessarily maximum hardness) and those that are economically acceptable.

### 3. Tool materials for machining nickel-based alloys

Of the many variables affecting any machining operation, the cutting tool, although small and relatively inexpensive, is one of the most critical. The requirements for any cutting tool material used for machining nickel-based alloys should include [3]: (i) good wear resistance; (ii) high strength and toughness; (iii) high hot hardness; (iv) good thermal shock properties; and (v) adequate chemical stability at elevated temperatures.

High-speed steels and cemented-carbide cutting tools are widely used for machining nickel-based alloys and these tool materials were the only choice for the machining of exotic superalloys for several decades. High speed steels are usually employed for intermittent cutting operations (milling, drilling, tapping and broaching, etc.), whilst cemented carbides are mainly used for continuous cutting operations (turning, boring and facing). Currently, some new ceramic tool materials (such as  $\text{Al}_2\text{O}_3$ -TiC mixed ceramics,  $\text{Si}_3\text{N}_4$  ceramics/Sialon and the latest SiC whisker-reinforced  $\text{Al}_2\text{O}_3$  ceramics (containing  $\sim 25\%$  SiC whiskers) have been used increasingly for machining nickel-based alloys [4]. The

use of multi-layer (TiN + TiCN + TiN) coated carbide tools, produced by the physical vapour deposition (PVD) technique have also shown remarkable improvement in the machining of nickel-based alloys [7,8]. Cubic boron nitride (CBN) cutting tools appear to give better overall performance than cemented carbides when machining nickel-based alloys [9], although the principal application areas for CBN are in the machining of ferrous materials [10]. The cost of CBN tools do however place a strict limitation on their use.

#### 3.1. Machining with carbide tools

The recommended cutting speeds range from 10 to 30  $\text{m min}^{-1}$  when machining nickel-based alloys with cemented carbide tools [1]. Severe flank wear and notching at the tool nose and/or the depth of cut line are the dominant failure modes when machining with carbide tools [7,8,11]. Cemented carbide tools cannot be used to machine nickel-based alloys at high speed since they cannot withstand the conditions of extreme high temperature and stress in the cutting zone. Rapid increase in notching occurs on carbide tools at higher cutting speed. This usually leads to the premature fracture of the entire insert edge [12]. Premature fracture of cutting tools can be avoided by employing the taper turning (or ramping) technique where the depth of cut is gradually shifted during machining, thus shifting the notch wear along the entire flank face of the tool. This will consequently lead to the generation of a uniform flank wear on the cutting edge. Recent advances in coating technology have led to the introduction of multiple coatings on cemented carbide cutting tools which tend to increase the achievable cutting speed up

to  $50 \text{ m min}^{-1}$  and beyond in single-point turning operation.

Lower cutting speeds are usually employed when milling (interrupted cutting) nickel-based alloys with carbide cutting tools. Chipping and/or fracture of the tool edges are the dominant failure modes due to a combination of high temperature, thermal and mechanical shock/stress, as well as the adhesion of the work material on the tool rake face [13–15]. K20 grade of cemented carbide has been found to give optimum performance when milling nickel-based alloys (Nimonic 75 and Inconel 718) at various cutting conditions [11,16].

### 3.2. Machining with ceramic tools

Pure oxide ( $\text{Al}_2\text{O}_3 + \text{ZrO}_2$ ) ceramic tools have not been effective in machining nickel-based alloys despite their good chemical stability, high hot hardness and high wear resistance. The poor performance of pure oxide ceramic tools can be attributed to their poor thermal shock resistance, low fracture toughness or low resistance to mechanical shock at elevated temperature. Single-point continuous turning tests on Inconel 718 showed that the mixed ( $\text{Al}_2\text{O}_3 + \text{TiC}$ ) ceramics outperformed the pure oxide [17]. The use of mixed ceramic tools was first reported in the late 1970s, with achievable cutting speeds of between  $120$  and  $240 \text{ m min}^{-1}$ , almost ten times faster than those used for plain carbides [18]. Tool life under these conditions was severely limited by excessive notching at the depth of cut region via welding and pull-out, which may be caused by the relatively low mechanical toughness of ceramic tools [19]. Reasonable tool lives have been recorded when machining Incoloy 901 with mixed ceramic tools at high cutting speeds up to  $300 \text{ m min}^{-1}$  [20,21]. The high speed machining of nickel-based alloy (Inconel 718) has also been achieved with another grade ( $\text{Al} + \text{Zr} + \text{W}$ ) of mixed ceramics [22,23].

SiC whisker-reinforced alumina ceramic tools were developed especially for machining nickel-based alloys [24]. They have also shown encouraging results when machining hardened steels and cast iron [25,26]. High cutting speeds ( $200\text{--}750 \text{ m min}^{-1}$ ) and high feeds ( $0.125\text{--}0.375 \text{ mm rev}^{-1}$ ) can be achieved when machining nickel-based alloys with whisker reinforced alumina ceramic tools [4,21]. Machining trials on nickel-based alloys showed that whisker-reinforced alumina ceramic tools were more favourable for machining the nickel-based alloy Inconel 718 in dry and/or wet cutting conditions [21,26,28].

The relatively high toughness of sialon ceramic tools also enables them to machine superalloys, hardened steels and cast iron at high speeds and high feed rates [29–32]. Up to 70% reduction in the rate of notch formation has been reported when machining Incoloy

901 with silicon nitride based tools. This improved performance gave a 57% cost saving in comparison with plain carbide tools in the semi-finish turning of Inconel 718 [31]. It has also been reported that sialon tools outperform SiC whisker-reinforced tools in the machining of Incoloy 901 [27]. One recommendation is that sialon ceramic tools perform better in machining the nickel-based alloy Waspaloy in the presence of coolant [21].

### 3.3. Machining with CBN tools

Turning tests on Inconel 718 using CBN tools with 30–95% CBN contents showed that notching decreased significantly with increasing CBN content, while average flank wear showed little variation, reaching a minimum at  $\sim 55\%$  CBN content [33]. Pressure welding of the work material onto the tool cutting edge was also observed: the degree of the welding was found to increase with decrease in the CBN content, thus increasing notching. The high notch wear resistance of the high-CBN-content tools make them suitable for machining nickel-based alloys [34,35]. The superior performance of the high CBN content tools is due to their higher hardness, since their hardness increases almost linearly with increases in the CBN content [36,37].

## 4. Surface integrity when machining nickel-based alloys

It has been reported that nickel-based alloys strengthened by heat treatment are very sensitive to microstructure change due to their high strength at high temperature, high ductility, high tendency to work hardening, etc. [38]. Major changes in the machined surface layer may include: (i) residual stresses (tensile or compressive) induced in machining; (ii) changes in hardness of the surface layer due to work hardening; (iii) microcracking or macrocracking, particularly in grinding; (iv) tears, laps and crevice like defects associated with the ‘built-up edge’ produced in machining and severe flank wear; (v) plastic deformation as a result of hot or cold work; (vi) metallurgical transformations attributed to high temperature; and (vii) chemical change including high temperature oxygenation and diffusion action between the workpiece and tool materials.

From some grinding tests on nickel-based alloys [2,3], varying amounts of residual stress are induced in the machined surface. The tests show that the stress levels on a machined workpiece surface vary with different metal removal rates and the residual stress may be either tensile or compressive. For single point turning, the use of a sharp tool with a zero wear land and positive rake face tends to minimise the amounts of residual stress, plastic deformation and work hardening

[3,17]. The shape of the cutting tool can also affect the machined surfaces. Improved surface finish and minimum surface damage can be obtained when machining with round inserts. Prolonged machining tends to increase the hardness of the surface layer and also deteriorates the machined surfaces [16,17]. This can be attributed to severe flank wear, and consequently increased component forces and cutting temperature due to the increased contact area and relative motion between the flank land of the tool nose region and the freshly machined surface of the workpiece. It has also been reported that significant tearing and considerable microstructural changes occurred on the machined surface of the nickel-based alloy, Inconel 718 [8,17]. Machining conditions that introduce high tensile strength on the machined surfaces should always be avoided and efforts should be made to ensure a compressive stress wherever possible. Undesirable effects in the machined surfaces can be minimised by the use of proper machining methods and machining conditions.

## 5. Tool failure modes and wear mechanisms when machining nickel-based alloys

### 5.1. Tool failure modes and wear mechanisms

Cutting tool materials generally undergo severe thermal and mechanical changes when machining nickel-based alloys. The applied stresses and temperatures generated at and/or close to the cutting edge greatly influence the wear rate and hence, the tool life. Notching at the tool nose and/or depth of cutting regions is the prominent failure mode when machining nickel-based alloys. This is caused by combination of high temperature, high workpiece strength, work hardening, abrasive chips, etc. [4,7,8,21,27]. Flank wear, chipping and catastrophic failure are other causes of tool rejection when machining nickel-based alloys.

Much work has been carried out to investigate notch formation and a number of theories have been proposed. The special physical conditions such as steep stress and temperature gradients have been suspected as probable causes. The heavily work-hardened surface being machined has been considered as a probable cause. No consensus of opinion exists as to the cause of notching when machining nickel alloys. However, a common agreement is that notch formation results from a combination of many factors and not necessarily from a single wear mechanism. It has been reported that a large temperature gradient has little effect on depth of cut notching [39], thus leaving the work-hardened layer as the main cause of notch formation during machining [16,21,40]. The prevailing theory is that notching is a result of chip–tool welding and consequent pull-out of tool materials [4,40,41]. The work

hardened layer often results in burr formation during machining. Further machining in the presence of significant burr usually leads to catastrophic fracture of the entire insert edge.

Shaw et al. [41] reported that the chip had red hot edges and a dark, cooler centre when machining nickel-based alloy (Waspaloy). They concluded that the edge of the chip, which is not restricted, receives a greater specific energy input than the centre, and which induces large welds and heavy pull-out of the tool materials at the edges. However, this anomalous phenomenon was not observed in later work by Khamsehshadeh, who separated the possible ways for notching into two processes [21]. The chip–workpiece separation process occurs with extensive lateral plastic flow of workpiece material at the edge of the shear zone. This first causes side flow to form a burr and then side flow on the chip which causes the chip to spread, both of which occur cyclically. Lee et al. [38] also reported that the high strength of nickel-based alloys causes high temperature and stress in the tool–chip contact area and that the separation of the edge of the chip from the workpiece is largely a tearing process. The fatigue loading on the tool, the work-hardened layer and the adherence of work material on the notched area and subsequent dislodgement, contributed to notching wear [21]. Notching has also been attributed to diffusion–attrition wear mechanisms [21,42].

It is interesting to note that severe notching of cutting tools occurs when machining nickel-based alloys under low speed conditions. The wear rate of ceramic tools when machining under high speed conditions is often not as great as may be expected under lower speed conditions [43]. This is supported by turning tests on Incoloy 901 [37] and may be attributed to the softening of the workpiece. Another advantage of machining nickel-based alloys under high speed conditions is chip control. Tough continuous stringy chips are usually produced when machining under lower speed conditions. These tend to change to small segmented chips when the cutting speed is increased to  $> 120 \text{ m min}^{-1}$  [44].

Although the exact details of various investigations on notch formation during machining do not agree, their principles are generally similar. In order to minimise notching at the depth of cut region, it is recommended to use a cutting tool with a large side cutting edge angle and negative rake. A recent investigation has demonstrated how notching of the cutting tool can be suppressed or even eliminated by employing recently developed multi-layer coated carbide tools as well as by using the taper-turning (or ramping) technique where the depth of cut is continually varied along the length of cut [8,12]. The temperature distribution in cutting tools used to machine nickel-based alloys has been studied [12]. The outstanding feature in the machining

of nickel-based alloys is that tool temperatures are far higher than in conventional steels due to their high strength, low thermal conductivity, etc. Furthermore, the tip of the cutting edge was frequently the hottest region when machining nickel-based alloys, while the maximum temperature when machining steels was always on the rake face some distance away from the cutting edge.

It has been observed that cemented carbide tools used for machining nickel-based alloys at speeds  $> 30 \text{ m min}^{-1}$  failed due to thermal softening of the cobalt binder phase and the subsequent plastic deformation of the cutting edge [45]. Examination of worn tools revealed a layer of 'disturbed material' beneath the crater and the cutting edge [46]. In this layer, the tungsten carbide grains were smaller and more rounded than in the original material. A correlation between the location of crater wear and the site of maximum temperature exists to support these observations. The common crater location arising from steel machining is on the rake face away from the cutting edge. Turning tests on Inconel 718 with a variety of tool materials showed that the crater started immediately behind the cutting edge [46]. The crater profile changed as the cutting speed increased, the deepest point moving closer to the cutting edge, where the highest temperature are developed.

It has been reported that diffusion or solution wear limits tool performance when machining Inconel 718 with cemented carbide tools [45]. This view is supported by Ezugwu et al. [7,11,12,15]. Diffusion wear resulting in a smooth surface was also observed when machining nickel-based alloys with ceramic tools [21,47]. The wear rate depends largely on the cutting temperatures and the chemical composition of both the tool and the work material. Machining at high speeds will accelerate the diffusion process due to higher temperatures. Bhat-tacharyya and Jawaid [48] reported a diffusion mechanism when machining Incoloy 901 with Sialon tools. The titanium, chromium, nickel and iron from the work material reacted with aluminium, silicon, and yttrium from the tool material. They claimed that Ti and Cr can have a stronger effect than Ni and Fe. These elements diffuse into the glassy phase of the tool matrix and reduce its strength. Tool failures caused by flank wear and crater wear may partly be attributed to the diffusion wear mechanism.

Chipping and/or fracture of the tool edge are the dominant failure modes when milling nickel-based alloys (Inconel 718 and Nimonic 75) [11,16,47]. Attrition wear and thermal cracking at high temperature were found to be responsible for tool fracture. Attrition wear is the removal of individual or aggregates of grains of tool material by the work material which subsequently leaves a rough area. This type of wear has been reported by many researchers when machining nickel-based alloys [21,47,49]. Attrition wear can be attributed

to irregular flow of the work material over the cutting edge of the tool, fatigue induced by the serrated chip, the formation of cracks generated by thermal and/or mechanical fatigue, etc. Significant plucking of tool particles when milling Nimonic 75 can be reasonably associated with attrition wear [16]. Breakdown of the cutting edge of cutting tools was also attributed to the higher stresses applied immediately behind the cutting edge when machining nickel-based alloys because the normal stresses on the tool are roughly twice as high for machining nickel-based alloys as for machining steel using the same cutting parameters [50]. The stress concentration at the cutting edge tends to deform the tool edge plastically, thereby encouraging tool wear by chipping, especially in interrupted cutting operations [17]. The abrasion wear mechanism has also been reported when machining nickel-based alloys with ceramic tools [21,48]. The plucked-off particles from the tool substrate and hard abrasive carbide particles contained in the nickel-based alloy are mainly responsible for abrasion wear which accelerates flank and crater wear in conjunction with the diffusion wear mechanism.

## 5.2. Effect of cutting environment on tool lives

It has been reported that the machining of nickel-based alloys in the presence of argon and in nitrogen-rich atmospheres accelerated the rate of notch formation, while slower rate of notching took place in the presence of air and oxygen gases [21]. The newly generated surface during machining is highly reactive and may chemically combine rapidly with atmospheres. When machining in air and in an oxygen rich atmosphere, oxygen is readily available and this will rapidly form an oxide on the new surface. This new surface containing oxide tends to suppress potential bonding under these conditions. Cutting tool materials with high affinity for oxygen may become oxidised. These two processes will hinder notch formation by reducing the forces locally and longer tool lives will be achieved [21]. The high rate of notching when machining in the presence of argon and in nitrogen-rich atmospheres can therefore be attributed to the absence of oxygen in the vicinity of the cut and consequently higher bonding forces leading to easier 'pull-out'. Increase in tool lives has been reported when machining Incoloy 901 and Inconel 718 with sialon and whisker-reinforced  $\text{Al}_2\text{O}_3$  ceramic tools, respectively in the presence of oxygen rich atmosphere [21,48].

Coolant is one of the most influential factors affecting tool performance when machining nickel-based alloys [12,21]. The longer tool lives achieved when machining Waspaloy with sialon ceramic tools are attributed to the reduction of the overall temperature in the cutting area. The glassy phase ( $\beta$  prime grains) in the sialon tool matrix will start to soften  $> 1000^\circ\text{C}$ .

The application of the coolant will retard the removal of the  $\beta$  prime grains by preventing its softening. The improved performance observed when machining Incoloy 901 and Inconel 718 with SiC whisker reinforced alumina ceramic tools may also be attributed to the lower temperatures generated, which tends to reduce the thermal stresses arising from the differential expansion between the silicon carbide and alumina [21].

The presence of coolant during machining may reduce the process of fatigue by irregular contact of the hot chip on the cutting tool [21]. This will in turn reduce the cracks generated and limit the plucking action of the cutting tool during machining. Coolant may intermittently penetrate to the flank face and rake face of the tool and induce rapid temperature changes. Due to the presence of high cutting temperature and periodic penetration of the coolant to the surface in question, the cutting tool will be subjected to continuous expansion and contraction. A tool material with adequate thermal conductivity and a low coefficient of thermal expansion will therefore minimise thermal damage by minimising the temperature fluctuation at the cutting edge.

The use of a high-pressure coolant supply when machining nickel-based, Inconel 901, superalloy with cemented carbide tools gave lower tool lives than when machining with the conventional coolant supply [12]. The use of a high-pressure coolant supply results in a significant reduction in the tool–chip contact length, and hence in the contact area, which in turn increases the compressive stress at the tool edge with little change in the cutting forces. This will encourage the formation of notching, thus leading to a shorter tool life.

### 5.3. Effect of tool geometry and machining method on tool lives

The shape and geometry of the cutting tools play an important role in determining tool life and the nature of the machined surfaces. Rhomboid shaped ceramic tools generally gave poor performance when machining nickel-based alloys, whilst the round and square shaped inserts gave longer tool life as well as a better surface finish under similar cutting conditions [16,17,21]. This is due to the increase in included angles of round and square inserts which tends to increase the edge strength of the inserts and tool–chip contact area, resulting in lower stresses and temperature at cutting edge. The variation in the approach angle of the round, square and rhomboid shaped inserts also effects tool life when machining nickel-based alloys. Rhomboid shaped inserts with a greater approach angle tend to increase the chip curvature during machining, thereby enhancing chip segmentation. This will have an adverse effect on the fatigue loading at the depth of cut and ultimately on tool life.

High speed machining trials on Inconel 718 and Incoloy 901 alloys showed that notching on the tool materials can be suppressed by employing the taper-turning technique [27,46]. The taper-turning (or ramping) technique is achieved by the variation of the depth of cut during machining. This technique tends to constantly shift the depth of cut along the cutting edge, thus distributing the concentration of the expected notch wear. This technique also results in a significant reduction or complete elimination of fracture at the tool edge caused primarily by excessive notching at the tool nose and/or end of the depth of cut regions.

## 6. Conclusions

1. Factors governing the machinability of nickel-based alloys include tool life, limiting metal removal rate, cutting forces and power consumption, surface finish and integrity of machined component, chip shape/swarf disposal, high strength/hardness of the work material, etc. Of these, the most important considerations are the surface abuse of the machined workpiece and tool life.
2. Notching, primarily at the depth of cut region, is the dominant failure mode when machining nickel-based alloys with different cutting tool materials due to the work-hardened layer, chip–tool welding and consequent pull-out, attrition wear, etc. Flank wear and chipping/fracture are also tool-life-limiting factors caused by diffusion wear, abrasion wear, attrition wear, and mechanical and thermal fatigue loading on the cutting tools, acting singly or in combination.
3. The presence of different atmospheres produces different effects on the tool wear rate when machining nickel-based alloys. Notching decreases when machining in the presence of oxygen but becomes more severe when cutting in the presence of argon and nitrogen.
4. Coolant plays a very important role in the machining of nickel-based alloys. The application of coolant will reduce the overall temperature of the cutting area and the process of fatigue induced by irregular contact of the hot chip on the cutting tool. However, the use of high-pressure coolant supply to machine nickel-based alloys produces a shorter tool life despite the benefit of improved chip segmentation. The high-pressure coolant supply causes a significant reduction in the chip–tool contact length/area and consequently an increase in stresses at the cutting edge.
5. The machinability of nickel-based alloys can be affected by the cutting tool geometry. An increase in the included angle and/or the nose angle of the inserts will bring about an increase in tool strength

and tool–chip contact area as well as a reduction in the approach angle, resulting in longer tool lives. Round and square shaped inserts will therefore give better performance than rhomboid shaped inserts.

6. Modern ceramic cutting tools (mixed ceramics, sialon, SiC whisker-reinforced alumina ceramics), multi-layer coated carbide tools and CBN tools present good overall potential for the machining of nickel-based alloys. In terms of capital cost, CBN compacts are currently about two to ten times those of the equivalent sialon or SiC whisker-reinforced alumina and coated carbide tools. Modern ceramics and recently developed coated cemented carbide cutting tool materials are therefore ideal for the machining of nickel-based alloys. The inadequate fracture toughness of ceramic tools can be overcome by employing the taper-turning technique.

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