# A high performance wrought nickel-base superalloy EI-929

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EI-929, a high performance Soviet nickel-base wrought superalloy having the composition, Ni–14Co–10Cr–5W–4Mo–4Al–2Ti–0.5V–0.1C–0.02B (wt %) was prepared by vacuum induction melting and vacuum arc refining. The ingots were homogenized and hammer forged. The microstructural inhomogeneities in the forged product were examined. Tensile and stress–rupture properties, and fracture behaviour were observed to vary between the central and the outer portion of the forged stock.

## 1. Introduction

The mechanical properties of wrought nickel-based superalloys are expected to depend upon their chemistry, processing and thermomechanical treatments. EI-929, with a nominal composition Ni–14Co–10Cr–5W–4Mo–4Al–2Ti–0.5V–0.1C–0.02B (wt %), is a nickel-base wrought superalloy which finds applications as rotor blades in advanced Soviet aeroengines [1–4]. The short term as well as long term mechanical properties of EI-929 are comparable to the Western alloy Nimonic-105, though the former has a lower proportion of expensive elements such as chromium and cobalt. No information is available about the effect of processing parameters on the mechanical properties of this Soviet alloy. Being a wrought superalloy, EI-929 would be extruded/forged to produce the desired components. In the forging, the microstructure is expected to differ in the various regions due to many different effects such as temperature inhomogeneity, chilling due to anvil/hammer, barrelling on the outside surface etc., thereby causing a variation in the resultant mechanical properties.

The aim of the present investigation was to prepare this alloy by different processing routes, hammer forge it, and evaluate the mechanical properties after a suitable heat treatment. The temperature dependence of the tensile and stress–rupture properties has been examined for the wrought alloy prepared by vacuum induction melting (VIM) and VIM plus vacuum arc refining (VAR) techniques. Microstructural and mechanical property anisotropy has also been investigated by examining specimens taken from various regions of the forgings. Specimens both parallel (L) and transverse (T) to the forging direction were examined.

## 2. Experimental procedure

### 2.1. Melting and casting

Primary melting of the alloy was performed in a 20 kg vacuum induction furnace under a vacuum of 0.7 Pa. The charge consisted of elemental additions of nickel, chromium, cobalt, aluminium, titanium and carbon, and master alloys of molybdenum, tungsten, vanadium and boron.

Ingots (75 mm in diameter and 400 mm long) were vacuum cast, skinned to remove the surface defects, radiographed and cropped. Some of the ingots were further vacuum arc refined using a 110 mm diameter copper crucible.

### 2.2. Hot forging

Heat treatment and metallographic studies were carried out on both the VIM and VAR ingot samples to determine the $\gamma'$ solvus, incipient melting and homogenization temperatures. Soaking periods were varied from 58 to 175 ksec at 1425 and 1455 K for homogenization. Hot torsion experiments were conducted on the specimens machined from the investment cast rods to select an optimum hot forging schedule.

The VIM and VAR ingots of 75 to 100 mm diameter and 75 to 100 mm length were forged by a 1000 kg hammer after being soaked at 1455 K for 7 ksec. Initially, three to four light blows were given followed by three to four medium-heavy blows with intermediate soakings at 1455 K for 0.6 ksec. This cycle was repeated five to six times to obtain a deformation of 40 to 50 per cent. The forged disc was machined off to remove the surface cracks and other irregularities. A number of 50 mm long and 10 mm square cross-section bars were then cut (in directions both longitudinal and transverse to the forging directions).

TABLE I Chemical composition of EI-929

| Element | VIM alloy (wt %) | VAR alloy (wt %) |
|---|---|---|
| Ni | 58.0–59.5 | 58.0–59.0 |
| Co | 13.8–14.4 | 13.9–14.2 |
| Cr | 9.6–10.5 | 9.6–10.5 |
| W | 5.5–6.2 | 5.5–6.3 |
| Mo | 3.2–3.9 | 3.2–3.9 |
| Al | 3.5–4.1 | 3.2–3.7 |
| Ti | 1.9–2.4 | 1.7–2.0 |
| Fe | 1.6–2.0 | 1.6–2.1 |
| V | 0.3–0.6 | 0.3–0.6 |
| C | 0.06–0.08 | 0.05–0.08 |
| B | 0.012–0.015 | 0.01–0.011 |
| Si | 0.11–0.17 | 0.12–0.15 |
| Mn | 0.01 | 0.01 |
| S | 0.003–0.008 | – |
| P | 0.01 | – |
| O | 22 ppm | – |
| N | 32 ppm | – |

## 3. Heat treatment and properties evaluation

The bars cut from the forged alloy were heat treated in air at 1220° C for 7 ksec followed by air cooling plus heat treatment at 1050° C for 4 h followed by air cooling [4]. Tensile specimens (gauge diameter 3.3 mm and gauge length 24.5 mm) and stress-rupture specimens (gauge diameter 3.30 mm and gauge length 20.0 mm) were machined from the heat-treated bars. The resulting specimens were then subjected to a 29 ksec ageing treatment at 1125 K followed by air cooling. Tensile tests at temperatures different from room temperature to 1275 K were conducted in air at a crosshead speed of $2 \times 10^{-2}$ mm sec$^{-1}$. A few specimens were also tested in vacuum (26 mPa) at temperatures of 1075 K and above. The stress–rupture tests were carried out in air.

## 4. Results and discussions

### 4.1. Chemistry

Table I shows the chemical composition of the VIM and VAR alloys and these agree with the reported chemistry [1–4]. The VAR ingots were generally observed to contain less carbon, boron, titanium and aluminium as compared to VIM, possibly due to oxidation related losses.

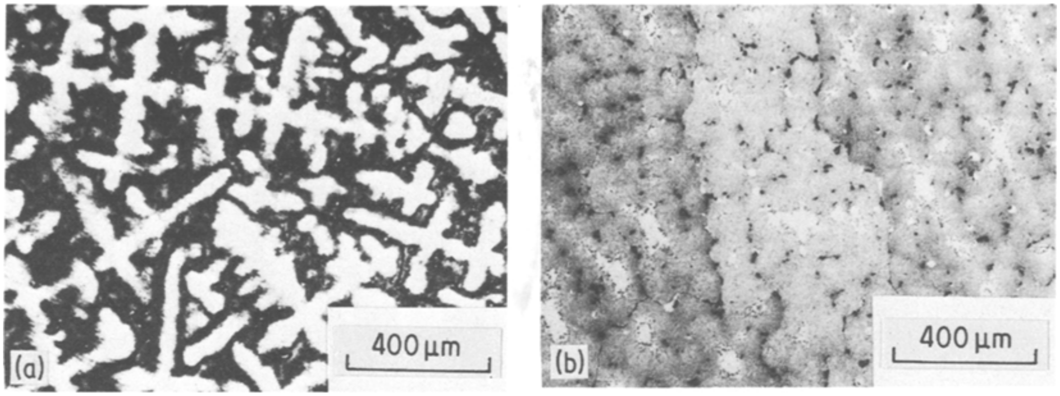
### 4.2. Microstructure

The microstructure of a VAR alloy sample is shown in Fig. 1a. The microstructures of VIM and VAR samples appeared similar except for the well known alignment of the grains due to some degree of directional solidification during the VAR process. The dendritic pattern and the resultant coring is quite evident in this microstructure.

The microstructure after homogenization at different temperatures is shown in Figs. 1b and c. A 58 ksec homogenization at 1455 K showed good reduction in the elemental segregation due to coring and also some partial dissolution of $\gamma'$ (white patches). MC-type carbides decorating the interdendritic and intergranular regions may also be seen in Fig. 1b. A 7 ksec homogenization at 1495 K resulted in complete removal of the coring effect (Fig. 1c) though $\gamma'$ still did not completely dissolve, suggesting that the $\gamma'$ solvus temperature is even above 1495 K. A 7 ksec exposure at 1525 K resulted in incipient melting. An 86 ksec homogenization treatment at 1455 K was therefore used in this study as it resulted in almost complete elimination of coring without severe loss of surface material due to oxidation.

### 4.3. Forging behaviour

EI-929 is expected to be a hard-to-work alloy owing to the presence of strong solid–solution and $\gamma'$ strengtheners like molybdenum and tungsten. Hot torsion experiments carried out on the samples made out of the investment cast rods showed that at 1415 K a maximum number of turns, i.e. 3.2, was obtained when the sample was subjected to a torque of 8.5 N m at a speed of 7 revolutions per second. Since a loss of temperature during transfer of the ingot from furnace to the forge hammer is expected, a soaking temperature of 1485 K was first selected for the



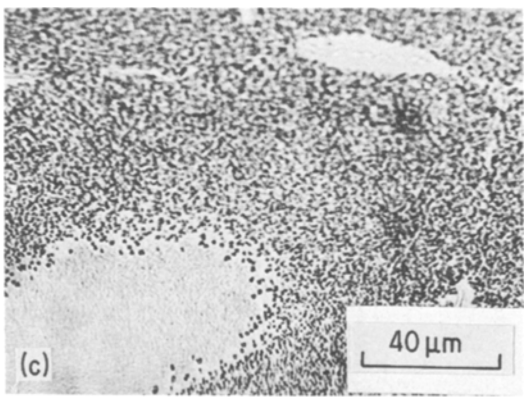




*Figure 1* Effect of homogenization treatment on the microstructure of VAR EI-929: (a) as-cast, (b) 1455 K/58 ksec, and (c) 1495 K/7 ksec.

hammer forging. However, it resulted in cracking without any significant deformation due to incipient melting. The temperature due to the heat of deformation must have risen from 1485 to 1525 K to cause this incipient melting. A reduced soaking temperature of 1455 K was therefore finally selected for forging of this alloy.

Region A (central portion) of the forged ingot (Fig. 2a) shows recrystallized equiaxed grains (Fig. 2b), region B shows a bimodal distribution of grain size (Fig. 2c), and region C shows the bimodal grain size distribution together with the original cast grain boundaries (Fig. 2d). The material in region A has undergone uniform deformation and microstructures of sections both parallel (SP) and transverse (ST) to the forging direction were observed to be similar. The material in region B has undergone non-uniform deformation, and therefore the grain structures on the ST and SP surfaces were not similar. The microstructure on the ST surface in region B was similar to that in region A (Fig. 2b) except for occasional large grains. Region C, on the other hand, has not undergone enough plastic deformation to break down the original cast structure completely. Therefore, the mechanical test samples with their gauge length parallel to the forging direction (L-type) were machined from region B, whereas, the specimens with their gauge length perpendicular to the forging direction (T-type) were machined from region A.

## 4.4. Effect of heat treatment on microstructure

The solutioning and two-step ageing treatments given to the forged alloy resulted in precipitation of about 40 vol% of secondary $\gamma'$ precipitates (Fig. 3a). The fine globular precipitates decorating the grain boundaries were possibly $M_6C$ or $M_{23}C_6$ types of carbides formed due to the ageing treatment. The fine carbide precipitates formed only on the grain boundaries and not on the twin boundaries. The $\gamma'$-precipitate size distribution was bimodal in nature (Fig. 3b). The larger MC-type carbide precipitates observed in the forged alloy, however, did not dissolve due to this heat treatment and remained distributed throughout the alloy.

## 4.5. Tensile properties

Table II lists the tensile properties of EI-929 alloy (processed by VIM and VAR routes) and these are almost similar to the values reported in the literature [4]. It is observed that the VAR alloy generally shows lower strength at room temperature as well as at higher temperatures as compared to the VIM alloy. This can be attributed to the lower titanium and aluminium content, as mentioned earlier, resulting in a decreased

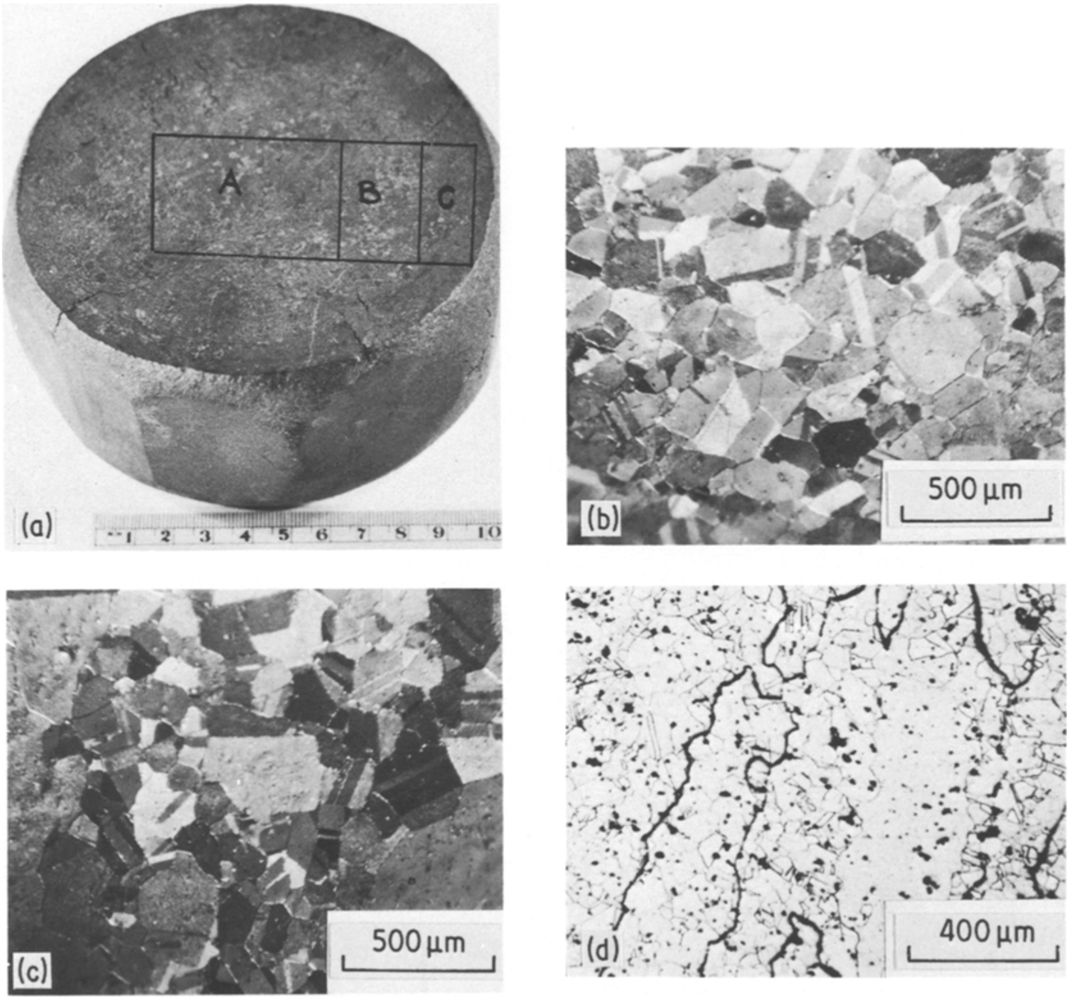



*Figure 2* Forging of a VIM EI-929 ingot. (a) Upset forged (soaked at 1455 K/7 ksec plus hammer forged). Microstructures after forging: (b) section transverse to the forging direction (region A), (c) section parallel to the forging direction (region B), and (d) section parallel to the forging direction (region C).

volume fraction of secondary $\gamma'$, and also decreased boron content in the VAR alloy. A comparison of the ultimate tensile strength (UTS), yield strength (YS) and elongation values between this alloy and Nimonic-105 [5] is illustrated in Fig. 4. EI-929 has a uniformly higher yield strength up to 1075 K. Specimens tested in vacuum gave a higher elongation as compared to specimens tested in air. This is possibly due to the elimination of the surface connected crack growth made faster by the oxidation. Tensile and yield strengths of T- and L-types of specimen are almost identical but L-type specimens generally have lower elongation values. This observation is true for both air-tested and vacuum-tested specimens.

The fractographic examination carried out on the fractured sections of both types of specimens showed that the fracture behaviour is intergranular for the T-type specimens tested at all temperatures from 875 to 1275 K (Fig. 5a). For the L-type specimen tested at 875 and 1075 K (Fig. 5b) the fracture is transgranular. However, when these specimens (L-type) were tested at higher temperatures, such as 1175 and 1275 K, the fracture was observed to be intergranular (Fig. 5c). The occurrence of transgranular fracture in L-type specimen at 875 and 1075 K is under-

TABLE II Tensile properties of heat treated EI-929 (L: longitudinal specimen from region B, Fig. 3a; T: transverse specimen from region A, Fig. 3a)

| Temperature (K) | Specimen condition | VIM | | | VAR | | | Estimated values from literature | | |
|---|---|---|---|---|---|---|---|---|---|---|
| | | UTS ($MN\,m^{-2}$) | YS ($MN\,m^{-2}$) | Elongation (%) | UTS ($MN\,m^{-2}$) | YS ($MN\,m^{-2}$) | Elongation (%) | UTS ($MN\,m^{-2}$) | YS ($MN\,m^{-2}$) | Elongation (%) |
| 300 | L | 1177 | 852 | 11 | 1107 | 765 | 7.5 | 980–1180 | 740–700 | 6–12 |
| | T | – | – | – | – | – | – | | | |
| 875 | L | 1004 | 858 | 4 | – | – | – | | | |
| | L | 998 | 829 | 4 | – | – | – | | | |
| | T | 1061 | 853 | 9 | – | – | – | | | |
| | T | 1098 | 856 | 9 | 866 | 724 | 9 | | | |
| 1075 | L | 901 | 817 | 4 | – | – | – | | | |
| | L* | 927 | 837 | 7 | – | – | – | | | |
| | T | 887 | 825 | 6 | – | – | – | 980 | | 6 |
| | T* | 972 | 827 | 12 | – | – | – | | | |
| 1175 | L* | 581 | 569 | 8 | – | – | – | 570 | | 7 |
| | L* | 569 | 541 | 7 | – | – | – | | | |
| | T | – | – | – | 476 | 415 | 5 | | | |
| 1275 | L* | 259 | 205 | 16 | – | – | – | 290 | | 15 |
| | T | – | – | – | 280 | 239 | 4 | | | |

*Tested in vacuum.

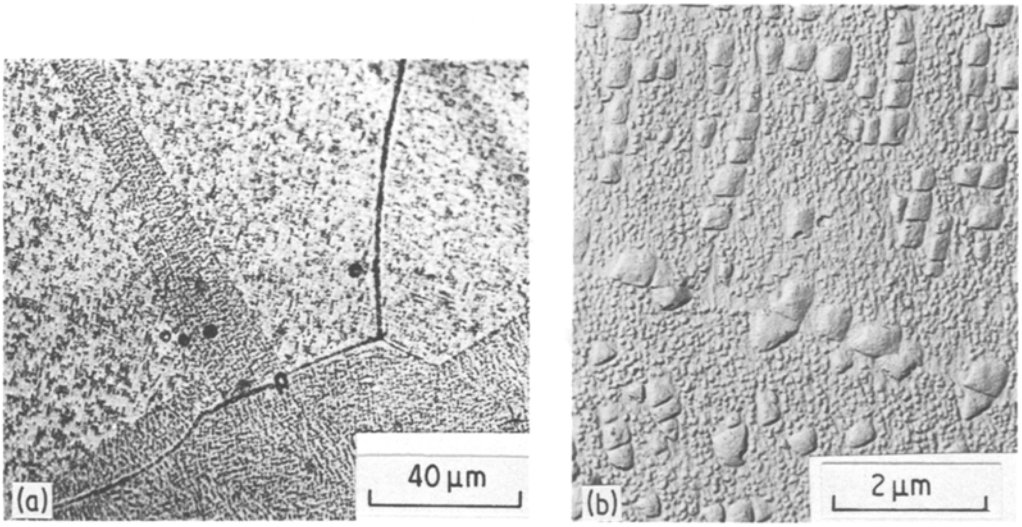



*Figure 3* Heat treatment of forged alloy. (a) Microstructure after solution treatment and two-step ageing (1495 K/7 ksec, AC + 1325 K/14 ksec, AC + 1125 K/29 ksec, AC). (b) Bimodal $\gamma'$ precipitation in the heat treated alloy. (AC = air cooling).

stood to be due to the alignment of the MC-type of carbide precipitates (Fig. 5d) helping propagate the fracture along their length and also getting fractured themselves in this process (Fig. 5b). Such alignment of carbide precipitates was not observed in T-type specimens taken from the uniformly deformed region A.

### 4.6. Stress–rupture properties

The rupture life of the EI-929 alloy specimens tested at various temperatures and stresses are listed in Table III, and the values measured generally agree with the reported data [4]. It is however seen that the VAR alloy, in general, is superior to the VIM alloy though the tensile properties of the VAR alloy were found to be lower than those of the VIM alloy. This is possibly due to the fact that the large chunky precipitates formed during the slow cooling involved in VIM ingot solidification are much more refined in size by the VAR operation. These large precipitates, mostly carbides, do not break down completely

TABLE III Stress–rupture properties of heat treated EI-929 (L: longitudinal specimen from region B, Fig. 3a; T: transverse specimens from region A, Fig. 3a)

| Temperature/stress ($K/MN\,m^{-2}$) | Specimen condition | Life (ksec) | | Estimated life from literature (ksec) |
|---|---|---|---|---|
| | | VIM | VAR | |
| 1075/333 | T | 2280 | 2510 | |
| | T | 2160 | 2370 | |
| | L | 540 | – | 1400 |
| | T | – | 3930 | |
| 1175/137 | T | 2190 | 3850 | 2200 |
| 1175/216 | T | 260 | – | 180 |
| 1175/245 | T | – | 300 | |
| | T | 130 | 140 | |
| | T | 148 | 210 | |
| | T | 110 | 250 | |
| | L | 104 | – | 144 |
| | L | 83 | – | |
| | L | 84 | – | |
| 1225/118 | T | 378 | 890 | 324 |

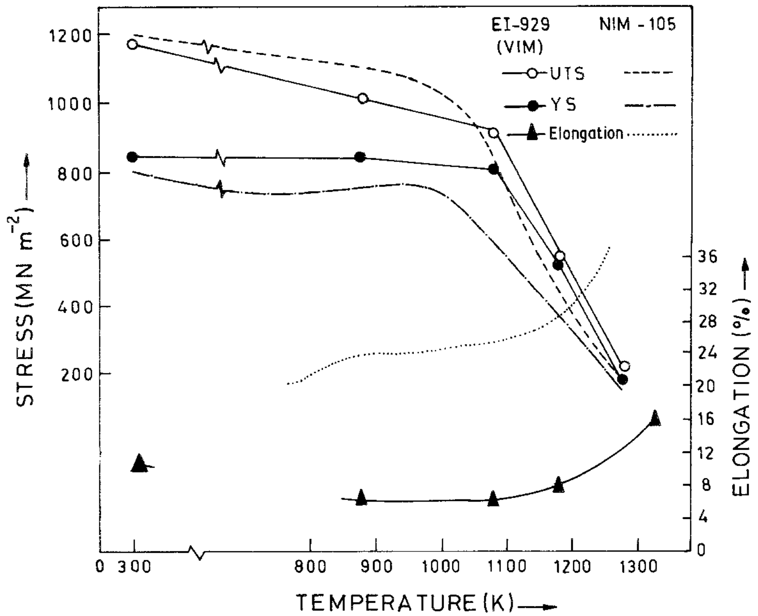



*Figure 4* Tensile properties comparison of EI-929 and Nimonic-105.

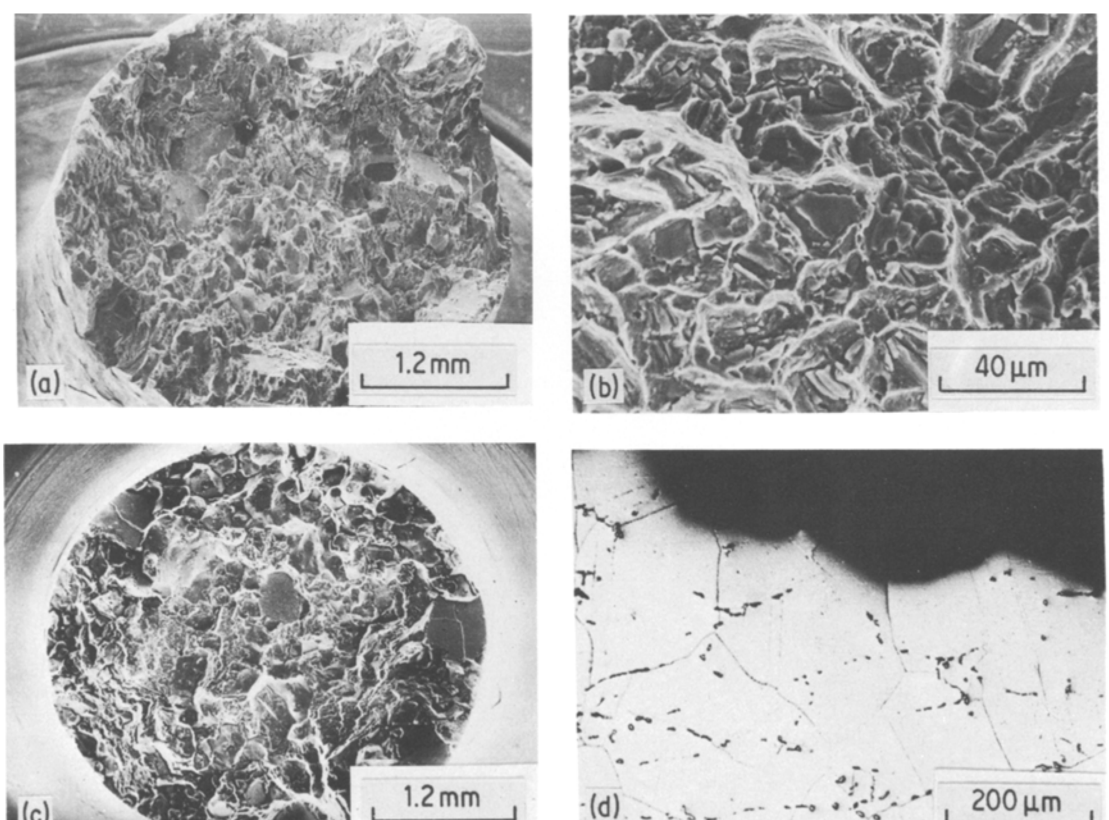



*Figure 5* Tensile fracture morphology for EI-929 specimens tested in vacuum. (a) T-type specimen, 1075 K; (b) L-type specimen, 1075 K; (c) L-type specimen, 1075 K; (d) L-type specimen, 875 K.

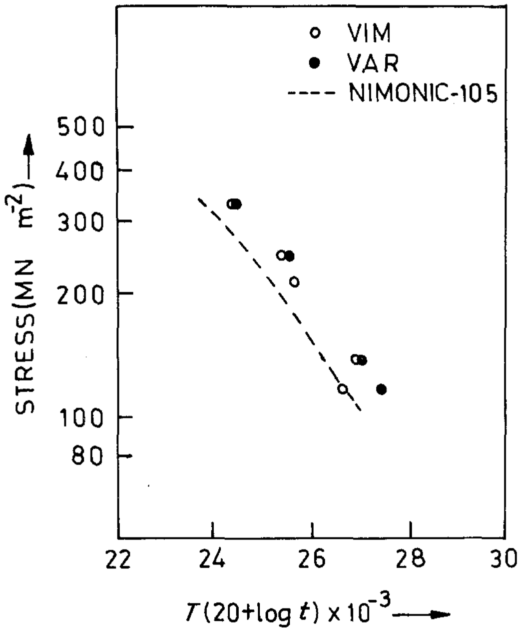



*Figure 6* Stress–rupture properties comparison of EI-929 and Nimonic-105.

even during the subsequent forging operation and must be providing the stress-raising sites for the rupture voids to nucleate. A comparison of the Larsen–Miller curves of this alloy and the Western alloy Nimonic-105 (Fig. 6) further confirms the superiority of the Soviet alloy with respect to stress–rupture properties.

The T-type specimens gave better stress–rupture life compared to the L-type specimens. In the case of short-time rupture tests (1175 K/ 245 MN $m^{-2}$), the rupture life of T-type specimens is generally longer than that of L-type specimens (Table III). This is confirmed from the long-time rupture test (1075 K/333 MN $m^{-2}$), where the life for the T-type specimen is about seven times that of L-type specimen.

## 5. Conclusions

Vacuum arc refined EI-929 possesses a higher stress to rupture life than does the vacuum induction melted alloy. Hammer forging of this high performance Soviet nickel-base alloy causes microstructural imhomogeneities in the forged stock. The material in the central portion of the forging has superior mechanical properties than that in the outer region, thus indicating probably that the alloy has to be first hot extruded in order to uniformly break the cast structure.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